

HEAT TREATMENT  
OF  
SOFT AND MEDIUM

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# HEAT TREAT OF SOFT AND MEDIUM

*THEORY AND PRACTICE OF THE PROPER  
TREATMENTS DESIGNED TO GIVE  
DUCTILITY TO STEELS USED FOR MACHINE*

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## TRANSLATOR'S PREFACE

It has been often observed that man progresses wasteful to the complex and economical. Progress is no exception. From direct iron and steel (in forges, and in one operation) we have come as commerce, made from ore through an intermediate product, pig iron.

Further progress may be expected along the demand for the maximum of strength in the inevitable, an end attainable by a further course either super-refining, perhaps in electric furnace carried far beyond the practices now commonly good quality.

Metallurgists and metallographists have only met with the fact that various impurities and additional properties of finished steel far in excess of that expected amount. Precise data along these lines are all therefore Dr. Giolitti's present book, containing discussion of their effect on commercial heat treatment, a powerful stimulus toward their study; proving as the advantages to be gained by their elimination or reduction. Respectful attention will be given by those who have seen the so-called "flakes" and "woody fractures" as understood and under control in the Italian work.

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# HEAT TREATMENT SOFT AND MEDIUM

## PART I

### THE PHENOMENA OF DIFFUSION IN P SOLUTION

#### CHAPTER I

#### GENERAL CHARACTERISTICS OF CRYSTALL AND MEDIUM STEELS AND OF ITS P HEAT TREATMENT

1. However extensive the series of mechanical treatments which may have been undergone during the life of a metallic alloy always retains a more or less evident memory of the characteristic properties directly resulting from the conditions of fusion and crystallization, those two stages of the formation of the alloy itself. Moreover, one can say that the majority of the various manufacturing operations performed upon metallic alloys—excepting those accomplished with cutting tools—have as their principal object the modification of these characteristics.

articles under treatment. As examples of this purpose, one may recall case-hardening of malleable iron, sherardizing, and a very few others.

Among other characteristic properties of steel in practice, one may mention the various defects of metal, and the kind, quantity, subdivision, and distribution of metallic inclusions, all of which are difficult to control in the solidified steel.

It is my purpose to study this first group of steel only indirectly at this time and only in so far as they exercise an influence which cannot be explained by crystallization phenomena, thus substantially excluding the first characteristic of the second group. As examples to be discussed may be mentioned (a) the distribution of a given element in a bath of steel on the extent of intracrystalline and extracrystalline during crystallization, and (b) the energetic distribution of metallic inclusions in certain cases upon the crystallization process.

3. A great deal less persistent are the properties of steel during the solidification process. In fact these properties depend partly directly and exclusively upon the crystallization. Now, this crystallinity is nearly always destroyed by means of operations on the totally solidified steel at a more or less high temperature, rolling either hot or cold, heat treating, etc.

There are of course some exceptions to this rule, which are frequent in general metallurgy and are extremely

(b) Chemical transformations in each crystal the "colonies" in which they congregate.

Modifications belonging to the first group can be brought about by mechanical or thermal manipulation, although the latter is more frequently depended upon. Modification of the second group, on the contrary, may be brought about almost entirely by means of heat treatment. Whenever a complex treatment includes one or more mechanical operations, the principal reason for the latter is usually to produce a desired crystalline form as to expedite and facilitate the work of the second group.

We shall later have occasion to see several examples of these modifications. My primary purpose in this volume is to investigate the conditions by which certain chemical modifications can be brought about in crystals in a steel. I shall mention the applications of these modifications only in those cases where its purposes and conditions are connected with those of the heat treatments. I shall therefore, to outline in a general way the broad scope of these transformations.

4. A concise statement may be made of the conditions under which the masses formed in a molten steel during slow cooling. The analysis of the process of solidification and of the structure of the resulting crystalline mass. Such reflection should be made immediately after the crystallization process has been completed. The metal consists of a conglomeration of mixed crystals of the various elements which enter into the composition of the steel in a solid solution in a iron.

the great majority of the steels used in practice are (iron, carbon, silicon, manganese, sulphur and nothing of those which contain nine or even ten per cent on the other hand, up to the present time completely have been studied only for alloys containing but

Nevertheless, from the standpoint of usefulness of theory the impossibility of making such a thorough research does not constitute an unsurmountable study of the problem. The reasons follow:

In view of the moderate concentrations of elements as they ordinarily occur in the large majority of steels in practice, it can be stated as a fact that the various elements segregate from the molten mass as soon as they solidify. *all belong to the same series*, or, in other words, all have similar characteristics. It should be remembered that we are dealing with steels covering a considerable range of alloying elements, those which contain the minimum concentration of each element for each different element or impurity in practice, and those which contain the maximum. Therefore we can be sure that there exists no discontinuity of solid solutions lying between the concentrations of elements in practice and pure iron.

Furthermore, excluding some very rare instances of importance, the mixed crystals appearing during the process of crystallization until the metal is totally solid are of the same family as those which formed when solid solutions are cooled, speaking in a physico-chemical sense rather than in a metallurgical sense. This statement remains true for many steels and

medium machinery steels, to which grades the process is limited.

5. Immediately after the steel freezes it is a solid solution of the various constituent elements of the metal—now totally solidified—cools further to a sufficiently low temperature, it passes successively through temperatures where  $\gamma$  iron is transformed into  $\beta$  iron,  $\beta$  iron into  $\alpha$  iron. When the intervals of metastable equilibrium in the lower region of each transformation range do not extend so low as to cause the allotropic transformation to occur at a velocity so small that it cannot be measured, the transformations actually occur at temperatures which depend upon the composition of the individual elements in solid solution in the steel and the velocity of cooling.

Now, among the various elements found in medium steels used in mechanical construction, carbon is the one which possesses the local concentrations which segregation may produce. It possesses the characteristic that its solubility in solid iron and of its compounds with iron or with the other elements varies with variations because of the allotropic transformations—variations so noticeable as to give rise to phenomena which are as to make it necessary to take them into consideration.

Consequently, the structural characteristics of steel cooled to ordinary temperatures depend upon the occurrence of two superposed orders of phenomena; namely, the formation of mixed crystals of one series and a later segregation of the entirely solidified metal caused by allotropic transformations.



almost never applied in practice; nor, if applied, would influence the properties of the metal to an extent comparable to that which the same treatments cause upon machine steels containing, say, less than 0.60 per cent. carbon.

The thoughts which I shall expound in the following pages will therefore apply only to this last grade of steels.

Here it is opportune to point out that in the extensive groups designated in metallurgy by the terms "dead soft," "soft," "medium" and "medium-hard" steels are also included steels applied to purposes very much different than for mechanical construction, but which, on account of their properties, are akin to those grades. As an example it is enough to remember that from a metallurgical standpoint we must include in the groups under discussion various types of steels used in the manufacture of projectiles, armor plates, etc.

6. Having thus briefly defined the type of crystalline structure possessed by the steels to which I intend to limit myself, it is possible to give in advance an outline of the fundamental characteristics of the heat treatments to be examined; that is to say, those heat treatments which result in permanent modifications in the constitution of crystalline structure and its chemical nature (see Sec. 3). And from what has already been said it is clear that these modifications can always be classified as one of the two following fundamental types:

1. Modifications obtained when subjecting the metal to temperatures higher than that of its last critical point on heating. Inasmuch as at such temperatures our steels consist of mixed crystals of only one series, it is evident that the only phenomena active can be nothing else but processes of diffusion of the various elements in solid solution in the iron, from the regions in which their concentrations are greater toward the surroundings where they are lower. This diffusion may be limited in sphere to one crystalline element, or may extend beyond to various parts of one or more crystalline colonies.

2. Modifications obtained by subjecting the metal to heat treatment in an interval of temperature within which falls one or more of its critical points. Such modifications consist essentially in changes of the state of segregation of the various phases which form in totally solidified metal as a consequence of allotropic transformation.

It is known that the character of the alterations of the second group is strictly dependent upon local variations in concentration of the numerous elements in solid solution in the iron. Therefore, modifications of the first classification to which the steel might have been subjected beforehand have a very strong action upon those of the second category.

7. From this brief summary it should be clear and definite that the operations I intend to study belong to those variously designated in

treatments" or "normalizing treatments" to distinguish them from the "final heat treatments" or "quality treatments," whose purpose is essentially different.

It is not easy to indicate briefly the characters which distinguish the treatments belonging to one group from those of the other. Perhaps one can say that normalizing heat treatments are those whose principal purpose is to modify the distribution of various elements which appear in solid solution, and also as a consequence to modify the distribution of the solid phases resulting from the decomposition of non-homogeneous mixed crystals at critical temperatures. The final state of aggregation of the equalized phases is in general immaterial to the successful completion of a treatment for homogeneity. In other words, even though the process of diffusion takes place through solid solutions which are metastable under the ordinary conditions of use for the manufactured article, yet the preliminary heat treatment may produce homogeneous metal existing either as austenite, martensite, troostite, or any of the other metarals to be found in the complete scale from the unaltered solid solution to the conglomerate finally produced by the complete separation of the various phases to which the solid solution itself gives place, and which are stable under the normal temperature of usage of the treated materials.

The determination of this exact position in the transformation series, on the contrary, is the principal purpose of the final heat treatments. Consequently such operations are intended to bring the metal into the form of a properly selected metaral (or metallic entity) chosen in relation to the mechanical properties desired in the treated metal.

In the case of soft and medium steels for machine construction the fundamental problem of the complete heat treatment is always the proper selection of the preliminary heat treatment. Contrary to the state of affairs for various other classes of steel, in these steels the final heat treatment usually does not present any special difficulties. In addition to this, the results which can finally be obtained depend largely upon the results of the preliminary treatments. A minute investigation of normalizing practice will not, therefore, seem inopportune.

In the discussion of these matters, in order to be brief and to the point, I will presume that the reader is acquainted with the general theoretical laws concerning the formation and transformation of the solid solutions in metallic alloys. I may be obliged at times to mention these fundamentals briefly, but only in order to present them under a form likely to simplify and clarify the conclusions which I shall deduce from them in relation to the special phenomena I propose to investigate.

Finally, it is now opportune to mention the fact that the purposes of this treatise are essentially of a practical nature. In fact I propose to explain, as completely as the meager scientific information available in this field permits, the mechanism of the physical and chemical phenomena

upon which the thermal processes used in practice are based, and to elaborate the subject in such a way as to make possible the rational application of these processes to a certain number of cases and also their eventual extension to cover special cases arising from time to time.

Once these purposes have been fixed, with a disclaimer of any intention to expound theories which shall not be immediately and directly applied to simplify and clarify the rationale of a technical process, it is clear that any considerations of a purely scientific character upon which my arguments must be based must be necessarily reduced to their simplest or diagrammatic representation. I must also omit all those rigid analyses and corollaries which should not be ignored in a complete and strictly theoretical treatise. Not finding an immediate and evident application to this study of technical operations, such exhaustiveness would only obscure the understanding of the nature of those phenomena.

Naturally I must admit that such a method, neglecting that which remains outside the small circle of a given group of phenomena, would be absolutely unacceptable in a treatise whose principal aim would be the study of the phenomena as such, the more so in the present inquiry, where accurate and complete theoretical knowledge is still very scarce—so scarce that elimination of any portion of that small stock can never be made without the chance of unconsciously altering the underlying principles even in their essentials.

It is necessary to say this in order to be absolutely clear as to the extent to which the conclusions I propose to develop may be generalized. However, due to the practical end in view, such necessary restrictions or limitations cannot present any great inconvenience. Always having care that well-proved facts are used as a basis, any eventual modification found necessary on account of a new weight assumed by some of the elements at first considered unimportant should not alter in any way the actual relations established by current experimental observations, and with the aid of this preliminary schematic but somewhat non-rigorous solution of the complex interrelations between the phenomena comprising technical processes. This is more than sufficient for the special practical ends I am aiming to reach.

Besides, whenever the student desires to dig deeper into the various scientific aspects of the subject, and to construct an exact balance between the sum of all the known experimental facts and the approximation to which I have arrived in their schematic interpretation, thus getting a precise notion of the real value of the latter, he can approach his task by reading the numerous existing treatises of scientific character, in which various features of the problem are thoroughly and rigorously studied, taking into account all available experimental data. To aid this inquiry a list of the most important books will be found at the end of the present

## CHAPTER II

### PRIMARY CRYSTALLIZATION OF A BINARY SOLUTION UNDER COMPLETE EQUILIBRIUM

8. As with all other liquids, solidification of molten steel of any composition starts at certain points, named "centers of crystallization" or "nuclei." Such tiny beginnings of the solid solution, in equilibrium with the melt at that temperature, contain a materially less concentration of the alloying elements present. Step by step as the cooling of the metallic mass proceeds, new layers of crystalline solid, each of slightly different composition, form around the primitive germs constituting the solid phase. At the same time new centers of crystallization originate in the residual liquid.

When solidification of the metallic mass is complete, the final arrangement of the crystalline structure is essentially determined by two factors:

1. By the position which the first crystallization centers assume and their influence upon the position of the subsequently forming centers. The spacing of nuclei largely governs the character of the crystalline colonies which constitute the steel and therefore controls the macrostructure of the metal.
2. By the laws governing the growth of the crystals as the temperature falls. These laws not only limit the boundaries of the resulting crystals, but more especially do they control the chemical composition of the successive crystalline layers formed around each pre-existent solid particle whose surface was in equilibrium with the residual liquid at a slightly higher temperature.

It is clear that the nature of the individual allotriomorphic crystals, and therefore the microstructure of the metal, depends primarily upon these laws of growth. Inasmuch as the effects of the phenomena of the first group depend, at least in part, upon the laws which regulate those of the second group, it is logical to start by studying the latter.

9. I shall begin, therefore, by analyzing the happenings at only one center of crystallization. This can always be done by assuming the existence of an impenetrable wall around the nucleus under consideration. A quantity of the metallic mass is thus isolated small enough to exclude the presence of all other centers of crystallization, either those centers forming at the start of crystallization or those

this procedure corresponds to considering an elementary portion of the totally solidified metal which is occupied exclusively by the solid layers which will have accumulated around the germ of crystallization formed at its approximate center. In this way one can isolate and study a "crystalline individual." I shall use this expression only in a chemical sense, and limit its meaning to the variation in composition in the various parts of the tiny mass considered, and exclude any reference to its crystallographic character. As a matter of fact it is known that usually such "chemical crystalline individual" does not correspond to a crystalline element or crystallite considered from a crystallographic point of view.

In order to represent these phenomena graphically, I shall use the familiar diagrams, where concentration is plotted against temperatures in rectangular Cartesian coordinates. As usual, temperatures are taken as ordinates along the  $Y$  axis (Fig. 1), and the concentrations of a given element  $X$  are plotted as abscissæ.

For a beginning, the effects of changing the concentration of any one of the various constituent elements in steel will be considered. In general, any one of these, such as carbon, manganese, sulphur, phosphorus, silicon, nickel, chromium, or others, may enter as a solid solution to a certain extent in commercially pure  $\gamma$  iron, and a discussion couched in general terms would apply with equal force to any of them.

Limiting ourselves to a binary system will greatly facilitate the preliminary diagrammatic representation of the course of crystallization, and in view of the before-mentioned uniformity and continuity of the mixed crystals constituting the metallic mass even after its solidification, this limitation will not jeopardize in any way the preliminary "qualitative" conclusions which may be deduced. "Quantitative" data concerning the phenomena of solidification and transformation cannot well be synthesized by the same simple method, because data for each one of the elements entering in the solid solution are modified numerically by the presence of each of the other elements. It would also be practically impossible to construct a diagram or model which would represent simultaneously the effects of all the elements entering into the constitution of the solid solutions under study. However, basing an argument upon the qualitative data obtained from the diagram showing the behavior of an isolated particle of the solid solutions under study, and taking into account the results of many direct experimental studies, both chemical and metallographic, of the various alloys used in practice, it is possible to reach even quantitative conclusions exact enough for the great ma-

the entire alloy, any abscissa  $x$  will represent a certain steel having  $x$  per cent. of element  $X$ .

It is well known that for each concentration of the element  $X$  lying within the range of analyses which are completely soluble in iron either in the solid or in the liquid state, there are two distinct critical temperatures related to solidificational processes. Of these, the first and higher corresponds to the first separation of mixed crystals from the molten mass; while the second and lower corresponds to the end of the crystallization, where the system is completely solidified. Normally, the temperature of the beginning of solidification cannot be lowered by a rapid chilling of the liquid metal; on the other hand, the temperature of complete solidi-

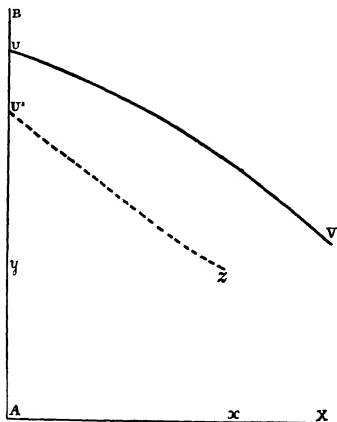


FIG. 1.—Solidification of complex solid solution.

fication is greatly influenced by the speed at which the immediately preceding cooling has taken place.

In Fig. 1, the point  $U$  represents the temperature of beginning of solidification of an alloy of iron and a constant amount of the other elements in their usual proportions with the exception of element  $X$ . For this alloy, the concentration of  $X$  is evidently zero, and its thermal events would be plotted on  $X = 0$ , or the  $Y$  axis.  $U'$  is the temperature of end of solidification after very slow cooling for the same alloy of iron with  $X = 0$ . Another steel containing a material amount of  $X$  has a corresponding temperature for the beginning of solidification, which can be

all such points of incipient solidification is  $UV$ , known to slope downward in case the solid solution separating out has a lower concentration in  $X$  than that in the molten mass. Since all the elements which enter into the composition of low carbon machine steels form solid solutions in iron with the just mentioned characteristic, this investigation can be restricted to the case where curve  $UV$  slopes downward with increase in the concentration of  $X$  in the alloy.

All the points in the same diagram representing the temperature of the end of crystallization of the various alloys with increasing concentrations of the element  $X$  form a curve  $U'Z$  located entirely underneath the first curve  $UV$ . As already observed, the point  $U'$  corresponds to the temperature of completed solidification of the alloy of iron containing a constant amount of the other elements in their usual proportions with the exception of the element  $X$ , which has a concentration of zero. Again it is true that this second curve has a downward slope when the concentration of the element  $X$  increases.

The position of the curve  $U'Z$ , as shall be seen, changes noticeably with variation in the velocity of cooling during crystallization of the metallic mass. We shall also see directly that these perturbations in the position of line  $U'Z$  are due to variations in the extent of the diffusion of each component throughout the mixed crystals and between them and the molten mass. Now, these variations in cooling velocity cast a simultaneous reflection in the amount of diffusion for all the elements alloyed with the iron as well as the element  $X$ . The same speed of cooling which even in the absence of the element  $X$  would cause the point  $U'$  to change a certain amount adds its influence to the displacement of  $U'$  caused by the element  $X$ , thus modifying the latter considerably.

For sake of simplicity we will not consider these last variations for the time being. This we can do, for example, by limiting the present discussion to the system in which only one element,  $X$ , is alloyed to pure iron. It is evident that in such case the two points  $U$  and  $U'$  will coincide in one, that one being the point of crystallization (and of fusion) of pure iron. This case is sketched in Fig. 2, and to this we shall now confine our attention.

11. Although, as already mentioned, the reader is considered to be familiar with the mechanism of the formation of solid solutions and of the subsequent segregation of the secondary constituents from them, nevertheless now I deem it advisable to make an exception to the plan of the book in presenting this first equilibrium diagram. It is the fundamental basis for the arguments which I propose to develop later. It should also be developed or synthesized in a way which will simplify and clarify further applications of these general principles to the various practical cases.

has a concentration  $D$ , and let us suppose it to have been heated to such a temperature that it is completely melted.  $D'$  then will be the point which represents the conditions as to temperature and composition existing throughout this molten mass.

It is well known that if the metal is allowed to cool until it reaches the temperature  $t'$ —corresponding to the intersection of the vertical  $DD'$  with the curve  $UV$ —mixed crystals start to separate from the liquid; in these crystals the concentration of the element  $X$  is that represented by the point  $F$ . For the reasons outlined above,  $F$  is situated at the left

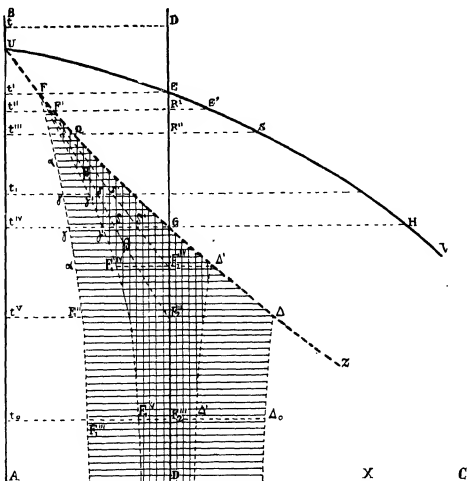


FIG. 2.—Solidification of binary solid solution.

of the vertical  $DD'$  in all the systems which the alloys comprising the subject of our study can present.

Solidification must start somewhere; and it is known that the first nuclei appear at determined points in the cooling melt. As I have already mentioned, the formation of these first germs of crystallization is one of the most important phenomena among the many occurring during solidification. The properties characteristic of the solid alloy depend to a very great degree upon the number of centers of crystallization and



treatment and mechanical work to which it will be afterwards subjected. I shall consider later these phenomena, which are dependent upon the sequence, propinquity and uniformity of distribution of the centers of crystallization, especially as influenced by the processes of undercooling. Only the process of crystallization around a single nucleus will now be studied, which is always theoretically possible, provided the mind is attuned to contemplate a metallic mass sufficiently small in dimension. We shall see later that, although it is not possible to cool a piece of metal in the laboratory so that only one crystal results, the conclusions which may be drawn from a theoretical consideration of this limiting simple case can be easily extended to the more complex practical cases, in which the metallic mass under observation actually crystallizes around a large number of centers.

12. As soon as the temperature of the metallic mass falls a very little below  $t'$ , reaching the value  $t''$ , the separation of a certain quantity of mixed crystals  $F$  will cause an increase in the concentration of the element  $X$  in the residual liquid, because the solid phase contains a smaller proportion of the element  $X$  than that contained in the liquid  $E$ . Let us call  $E'$  the new concentration of  $X$  in the residual liquid. Liquid  $E'$  is in equilibrium with solid of composition  $F'$  (greater than  $F$ ) at temperature  $t''$ . In other words, the first solid to separate from the molten liquid when the temperature falls the least bit below  $t''$  will have an analysis in  $X$  equal to an abscissa  $t''F'$ , richer in  $X$  than the nucleus about which it must precipitate.

From the definition of the curve  $UV$ , it is clear that the point  $E'$  will lie upon it.

If now we consider the conditions of our system containing homogeneous liquid and heterogeneous crystal after it has reached temperature  $t''$  and take due account of the fact that the cooling process in itself is essentially continuous, we easily see that this system cannot be considered as being in equilibrium from the standpoint of the phase rule.

In fact, should no other reaction occur the crystallite forming around the nucleus  $F$  will consist of a series of shells of differential thickness, in each of which, as you pass from the center outward, the concentration of  $X$  would vary from that value corresponding to the point  $F$  to that value corresponding to the point  $F'$ , showing a scale of all possible intermediate values between these two points. Diffusion must take place in such a compound crystallite to allow an approach to a state of true chemical equilibrium. In fact, diffusion of element  $X$  from a region of high concentration to one of low concentration does take place within the crystal of solid solution and between this and the liquid masses in contact with its surface, in a manner similar to those movements which render a liquid solution homogeneous, although with lesser velocity and

concentration in all parts of a continuous volume. It is also known that such a process—which theoretically could not end in a finite lapse of time—actually takes place in liquid solutions with such velocity as to produce substantially complete uniformity of concentration of substances dissolved in very large quantity of solvent in a relatively short period of time, and this although the initial concentrations vary greatly in regions of liquid quite remote one from the other. The linear velocity of diffusion in liquid solutions increases very rapidly with increase in temperature, until at very high temperatures it reaches a tremendous speed—as an instance, take the case of ferro-alloys shovelled through one door of an open-hearth furnace, but which seemingly spread throughout the entire bath very rapidly.

As has already been mentioned, the velocity of diffusion in solid solutions, although increasing very rapidly as the temperature increases, is much lower than in the case of a liquid solution. Consequently in annealing practice the effects are noticeable only between regions separated by very short distances—usually by a few tenths of a millimeter and but very seldom as great as a few millimeters.

This is true even at high temperatures, such as for instance  $800^{\circ}\text{C}$ . to  $1100^{\circ}\text{C}$ . ( $1470^{\circ}\text{F}$ . to  $2010^{\circ}\text{F}$ .) maintained for quite a long time, say 5 to 10 hours. Nevertheless we shall see later that even in solid solutions the processes of diffusion cause effects of the greatest industrial importance.

13. Returning to the thread of the argument, let us maintain our system at temperature  $t''$  for such a time that the composition of the solid particle is able to become practically uniform at all points. This condition will ensue as an effect of diffusion of the element  $X$  from the higher concentration periphery to the lower concentration center.

A somewhat more rigorous examination of this process would easily demonstrate that such equalization is necessarily accompanied by a gradual diffusion of the element  $X$  from the liquid where its concentration is  $E'$  into the external strata of the solid portion, where the concentration is being withdrawn to a figure slightly less than the abscissa  $t''F'$ .

Only when the process of diffusion has proceeded substantially to completion will we possess a system in the state of chemical equilibrium at the temperature  $t''$ . This system will be formed by a solid crystallite of uniform composition  $F'$  and of mass  $M'_s$  contained within a residual liquid of composition  $E'$  and of mass  $M'_l$ .

Now if a mass of solid  $M'_s$  has a concentration in  $X$  equal to the value  $t''F'$ , the total amount of  $X$  in the solid material is the mass times the concentration, or  $M'_s \times t''F'$ . Similarly the mass of liquid existing

$t''E'$ ; consequently the amount of  $X$  in the liquid is  $M_l \times t''E'$ . But the total mass of the system  $M_t = M'_s + M'_l$ . Since the average  $X$ -concentration of the whole system is  $\bar{t}''R'$ , the total amount of  $X$  existing equals  $M_t \times \bar{t}''R'$ . This is also represented by the sum of the  $X$  in solid plus  $X$  in liquid. In symbols

$$M_t \times \bar{t}''R' = M'_s \times \bar{t}''F' + M'_l \times \bar{t}''E'$$

Substituting for  $M_t$  its value  $M'_s + M'_l$ , multiplying and collecting.

$$\frac{M'_s}{M'_l} = \frac{\bar{t}''E' - \bar{t}''R'}{\bar{t}''R' - \bar{t}''F'} \quad \text{or} \quad \frac{M'_s}{M'_l} = \frac{\bar{R}'E'}{\bar{R}'F'} \quad (1)$$

The same phenomena analyzed above will proceed upon further cooling the system below temperature  $t''$ . Suppose that the cooling, although continuous, is slow enough so that complete uniformity of the concentrations of the element  $X$  is established at all times in each of the liquid and crystalline phases. Then when the temperature has fallen to lower temperature  $t'''$  our system will be formed of a quantity  $M''$  of mixed crystals of uniform  $X$ -concentration represented by the point  $Q$ , and of a quantity  $M''_l$  of a residual liquid, in which the concentration of the element  $X$  will have reached the value represented by the point  $S$ .

By symmetry it is known that the relation between the quantities  $M''_s$  and  $M''_l$  and the length of the two segments  $QR''$  and  $R''S$  is as follows when complete equilibrium exists:

$$\frac{M''_l}{M''_s} = \frac{\bar{QR}''}{\bar{R}''S} \quad (2)$$

or (if  $M_t$  represents the total mass of the system)

$$\frac{M''_l}{M_t - M''_l} = \frac{\bar{QR}''}{\bar{QS} - \bar{QR}''} \quad (3)$$

If the cooling is continued further under the same conditions, it is evident that crystallization must stop as soon as the quantity  $M_l$  of the residual liquid is reduced to zero. As can be readily seen from relation (3), this will happen when the length of the segment corresponding to  $QR''$  is reduced to zero—namely, at the point  $G$ .  $G$  is situated at the point of intersection between  $DD'$  and curve  $UZ$ .  $UG$  is the locus of all points such as  $F$ ,  $F'$ ,  $Q$  and  $G$ , representative of the concentrations of the element  $X$  in the mixed crystals at the various temperatures  $t'$ ,  $t''$ ,  $t'''$  and  $t^{IV}$  which, at those temperatures, are in equilibrium with the liquid phases whose compositions are represented by the conjugate points of the curve  $UV$  ( $E$ ,  $E'$ ,  $S$  and  $H$ , respectively).

It has thus been established that the intersection  $G$  represents the temperature and composition of a mixed crystal of equalized  $X$  concentra-

tion at the instant when it has absorbed the last trace of residual mother liquor in equilibrium with it. The composition of the latter was represented by the point  $H$ . Postulating a continuous progress of the cooling and crystallization, it is clear that the locus of all points analogous to the points  $F$ ,  $F'$ ,  $Q$  and  $G$  is a curve  $UZ$ . This curve shall be hereafter given its usual name, the "solidus." It coincides in the special case of continual diffusion to exact equilibrium with the curve representing the actual end of crystallization. The curve  $UV$  takes the name of "liquidus." It is the locus of all points representing the composition of the mother liquors in equilibrium with the mixed crystals whose composition is represented by abscissal intersections with the curve  $UZ$ . In the special case of complete equilibrium now under consideration, the liquidus coincides with the curve of the beginning of crystallization.

14. The validity of the conclusions which have been arrived at through the preceding pages is clearly dependent upon two principal conditions:

(a) It must be possible for the element  $X$  to enter into true solution with the iron both in the solid or liquid state up to all concentrations of that element found in the alloys here under consideration. Solubility must be possible in the mixed crystals as well as in the mother liquor in equilibrium with them—which, by the way, must carry a much higher percentage of element  $X$ .

(In other words: None of the illustrative points which have been considered in Fig. 2 may encounter a region where true solution cannot take place during all the process of crystallization. As I have already said, this condition actually obtains in all the chemical systems to which this study refers, for all the higher intervals of temperature to which we have now restricted our investigation. Therefore I shall reserve consideration of the occurrences when the temperature restriction is lifted, to the pages where we shall study the transformations taking place at temperatures below the end of crystallization.)

(b) At every instant during the solidification of the mixed crystals, the concentrations of the element  $X$  must be uniform in all regions of the crystalline mass then in existence, and uniform at all points of the residual liquid—although the concentration differs, naturally, between the solid and liquid phases reciprocally in contact and equilibrium.

Now, this second condition—especially as to the uniform concentration of  $X$  in the solid phase—never exists in the alloys which form the object of the present study, at least at the time and under normal conditions of solidification. In fact, the cooling velocity of the metallic mass throughout the entire interval of crystallization is ordinarily far greater than the snail's pace which would be necessary in order to allow the diffusion of the element  $X$  to take place in such measure as to render

in concentration. Intracrystalline heterogeneity is always present on first cooling, as I have already had occasion to remark, even when the dimensions of the crystallites formed around each crystallization-germ are extremely small, for instance, a few hundredths of a millimeter.

However, the opposite extreme never occurs in the cases we are now studying—the extreme where the processes of diffusion in the mixed crystals are practically stationary and zero.

Disregarding, therefore, this second extreme as well, we shall now start to analyze the process of crystallization as it really happens in industry; namely, those cases in which diffusion of the substances dissolved in the mixed crystals takes place to a moderate amount, yet at a velocity lower than that necessary to permit complete uniformity in composition of the cooled crystalline masses to be attained.

## ACTUAL COURSE OF CRYSTALLIZATION OF A LIQUID BINARY SYSTEM

15. Assuming that the last-mentioned hypothesis represents the actual facts, when the temperature of our system falls until it reaches  $t''$  (Fig. 2) a certain quantity of the element  $X$  will have diffused from the external layers of the crystalline mass where its concentration is  $F'$  toward the central nucleus of initial concentration  $F$ . At the same time new quantities of  $X$  pass into the external layers of the crystal from the liquid. The mother liquor will necessarily contain a larger percentage of  $X$  than that which should correspond to the conditions of complete equilibrium. This excess quantity of  $X$  is the same as the deficiency in the material surrounding the nucleus<sup>1</sup>  $F$ , i.e., the amount of  $X$  over (in the liquid) or under (in the solid) that amount which should correspond to equilibrium at this temperature.

In reality the diffusion in the liquid mass and between the liquid and the external layers of the solid solution crystals takes place at a much greater velocity than the process of diffusion between parts of the solid caused by the difference of concentration. Consequently for normal velocities of cooling the conjugate points determined by the same abscissa intersecting the liquidus curve  $UV$  and solidus  $UZ$  represent in reality and with a fair approximation the concentrations of the element  $X$  in the residual liquid and in the periphery of the mixed crystals in equilibrium with said liquid respectively.

In consequence, it is clear that once the system has reached the temperature  $t''$  it will be composed of a residual liquid in which the concentration of the element  $X$  will have the value  $E'$  and by a crystallized mass which will possess the composition corresponding to the point  $F'$  only in its external layers, while the underlying strata will contain gradually decreasing proportions of the element  $X$  layer to layer as one proceeds toward the center. Here it reaches a composition represented by a point  $F'_1$  (see Fig. 3, where, to make these phenomena more clear, this part of the diagram is reproduced to larger scale) situated between  $F'$  and  $F''$ . It will evidently be the nearer to the position  $F''$  the smaller has been the amount of diffusion of  $X$  in the solid solution.

It is, therefore, evident that the average composition of the total solid phase which has actually formed down to the temperature  $t''$ , will be represented by a point  $F'_2$  situated between  $F'_1$  and  $F'$ .

<sup>1</sup> "Nucleus" is used here in the sense of being almost a mathematical point.

the mass  $M'_{2i}$  of the residual liquid which is in equilibrium with it at the same temperature, is expressed in the following proportion similar to equation (1) (Sec. 13):

$$\frac{M'_{2s}}{M'_{2i}} = \frac{\overline{R'E'}}{\overline{F'_2R'}} = \frac{\overline{R'E'}}{\overline{R'F'} + \overline{F'_2F'}} \quad (4)$$

By comparing (4) with (1) it seems that the total mass of the solid phase actually separated at the temperature  $t''$  is smaller than that which

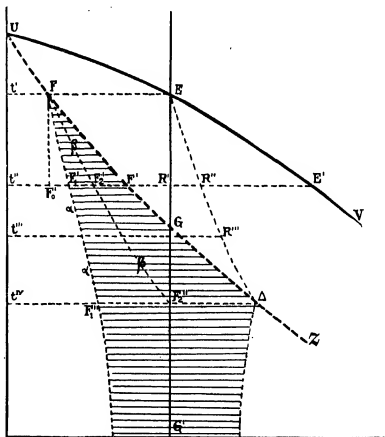


FIG. 3.—Crystallization of liquid binary system.

would have separated if diffusion took place in it sufficient to make the concentration of  $X$  totally homogeneous.

**16.** Continuing the cooling of the system, for each temperature two points may be determined corresponding respectively to  $F'_1$  and  $F'_2$ . From the definition of these two points as regards temperature  $t''$  it is clear that the locus of all points corresponding to  $F'_2$  will be a curve ( $F F'_2 F''_2$ ) passing through  $F$  and lying totally under and at the left of the solidus. This curve represents the average composition of the solid phase at the various temperatures, and is marked  $\beta$  in Fig. 3.

The locus of all the points corresponding to  $F'_1$  will be a curve  $F' F'_1 F''_1$ , also passing through  $F$ , and lying totally under and at the left side of

the preceding curve. This curve represents the minimum  $X$  concentrations in the solid phase at the various temperatures, or the  $X$ -concentrations at the crystalline nuclei at the various temperatures. This curve in Fig. 3 is indicated with  $\alpha$ .

As shown by equation (4) of Sec. 15, it is clear that the reacting mass will be totally solidified at the moment it reaches the temperature  $t^{\text{IV}}$ , or at  $F''_2$  where curve  $\beta$  intersects vertical  $EG$  representing the average analysis of the system.

At temperature  $t^{\text{IV}}$  our crystalline element will evidently have an average composition represented by the point  $F''_2$ . At its nucleus it will possess the composition represented by the point  $F''_1$ , corresponding to the intersection between the abscissa through  $F''_2$  with curve  $\alpha$ . Its external composition, however, will be that represented by the point  $\Delta$ , determined by the intersection of the solidus ( $UFG\Delta$ ) with the horizontal line drawn through  $F''_2$ .

Afterwards when the temperature of the system continues to fall below the temperature of total solidification,  $t^{\text{IV}}$ , diffusion of  $X$  between various parts of the crystalline element does not stop suddenly, but its effect continues to be felt, although with a rapidly decreasing intensity as the temperature of the system lowers. Consequently the concentration of the element  $X$  decreases at the periphery and increases at the center of the crystal. This fact, in our diagram, corresponds to a gradual converging of the points corresponding to  $F''_1$  and to  $\Delta$  toward the vertical  $EG$  as the temperature of the system falls.

Therefore, starting from  $t^{\text{IV}}$ , the temperature of the end of the crystallization, the diagram will have two descending curves—originating respectively at  $F''_1$  and at  $\Delta$ —marking out a band whose borders delimit the compositions of the solid phase at the edge and center of a single crystal. The width of this band (and the heterogeneity) decreases with falling temperature.

17. We could have reached exactly the same conclusion starting from the following observation:

Due to the fact that diffusion of  $X$  in the solid phase does not take place rapidly enough so that it could be considered as homogeneous at each instant, it follows that at a given stage in the freezing—for example, at temperature  $t''$ —the crystal previously formed contains a nucleus in which the concentration of element  $X$  is less than that which corresponds to complete equilibrium in the system. From the point of view of the equilibrium between the solid and the liquid phase, this is equivalent to a condition where a certain quantity of the element  $X$  has been added to those parts of the system which are actively engaged in the process of crystallization, namely, added to the metal forming the residual liquid and the outer surface layer of the crystal. Therefore,



following such a course as though the chemical composition of the system under transformation were represented by a point  $R''$  which may be called the fictitious composition of our system, situated of course to the right of the point  $R'$ . It is clear that the fictitious composition will constantly move to the right all the time the temperature is falling, occupying at temperature  $t'''$ , for instance, the position  $R'''$ . Thus, it will follow a curve  $E R'' R'''$ , cutting the solidus  $UZ$  at the point  $\Delta$ , the definition of which we had arrived at following another path.  $\Delta$  is situated below  $G$ , the point corresponding to the end of crystallization of the system in complete equilibrium. From the above explanation it follows directly that the points  $R''$ ,  $R'''$ , etc., will be moved further towards the right, and therefore the point  $\Delta$  will be also pushed further downwards with decreasing importance of the process of diffusion in the solid phase; in other words, the more rapid the cooling in the given system.

From the meaning of the curve  $UZ$ , it is clear that at temperature  $t^{IV}$ , corresponding to the point  $\Delta$ , the metallic mass will be completely solidified. Since  $F''_1$  represents also in this case the composition reached by the center of the crystalline element at the instant in which solidification of the mass ends, we find again the same band of variant composition in the solid field, bordered by two lines descending from  $F'_1$  and  $\Delta$ , exactly as has already been defined from another premise.

18. Let us see now what is the meaning of the points in the diagram, contained within this band. I will refer again to Fig. 2, page 13, where a larger heterogeneous area is represented ( $F''_1 \Delta F'''_1 \Delta_0$ ).

Let us consider the system at a temperature  $t_0$ , lower than the actual end of solidification ( $t^V$ ). Remembering the explicit limitation to the case in which the entire metallic mass has solidified around only one crystallization-germ, and thus is formed of only one crystalline element, we know by definition that the concentration of the element  $X$  at the nucleus of the crystalline mass will be represented by the point  $F'''_1$ , while the entire surface of the same mass will have an  $X$ -content corresponding to the point  $\Delta_0$ .

The successive layers of crystalline matter which build up around the nucleus during the successive steps in solidification evidently possess compositions in  $X$  represented by successive points along the line  $F'''_1 \Delta_0$  starting from the point  $F'''_1$ .

In other words, the successive points between  $F'''_1$  and  $\Delta_0$  represent the concentrations of the element  $X$  at similar points on any radius vector, ranging from the center to the surface of the crystalline element.

19. Now, let us again take the same system at a temperature  $t$ , at which it is totally liquefied, and let us suppose that it cools more slowly than it did previously.

In this case it is clear that the phenomena of diffusion will have opportunity to proceed more nearly to completion during each given temperature interval owing to the added time during which the system remains in it. Thus, for example, at the temperature  $t_1$  the concentration of the element  $X$  at the nucleus of the solid phase will be enriched to a value  $\gamma'$ , higher than the value  $\gamma$  which it had reached at the same temperature in the previous case. At the same time—and for the same reasons—the average  $X$ -concentration of the solid phase will reach a value  $\delta'$ , larger than the previous value  $\delta$ .

Repeating the same analysis for successively lower temperatures, the conclusion is reached that in this slower cooling the two curves  $\gamma'$  and  $\delta'$  will be displaced to the right of the respective positions ( $\alpha$  and  $\beta$ ) which they occupied during quicker cooling.  $\gamma'$ , as  $\alpha$ , represents  $X$ -concentration at the nucleus, and  $\delta'$  and  $\beta$  both denote the average composition of the solid phase with falling temperature.

As a consequence of slower cooling the temperature of the end of crystallization (determined by the intersection of curve  $\delta'$  with the vertical  $DD'$ ) will be higher than in the preceding case. At this temperature, the difference in composition between the nucleus ( $F_1^{IV}$ ) and the periphery ( $\Delta'$ ) will be smaller than before. Should further cooling be continued at a slower velocity than before, the width of the band of variant concentrations in the solid phase will narrow more rapidly. (See Sec. 18.)

Summing up, we can establish the following propositions which regulate the solidification of any given system capable of crystallizing in only one continuous series of mixed crystals:

*First.*—For each given velocity of cooling the temperature of the end of crystallization has a determined value. Other conditions being equal, this is lower the faster the cooling.

*Second.*—For each determined velocity of cooling the curve showing the composition of the crystalline nuclei and the curve showing the average composition of the solid phase assume two distinct positions. Other conditions being equal, the two curves are moved further to the left and downward, diverging more and more from the true solidus with the more rapid cooling.

In the extreme case in which the cooling is so rapid (by drastic quenching) as to entirely prevent diffusion during solidification it is clear that the curve representing the composition of the nuclei becomes a vertical line.

At the opposite extreme where cooling is so slow that at each step

the crystalline mass, it is clear that the two curves under discussion will both coincide with the solidus.

*Third.*—As a consequence of the above-indicated displacements of the two curves  $\gamma$  and  $\delta$ , the band representing the heterogeneity existing edge to center in the mixed crystals with slower cooling, is narrower at its top (that is, at the temperature of complete solidification) and continually becomes narrower at lower temperatures.

20. We will see later what great importance the width of this band has upon the structure and properties of steel, so that any treatment capable of producing modifications in it is capable of practical applications.

However, it is even now easy to see from the foregoing brief analysis of the circumstances to which it is due, that any process capable of modifying the characteristics of the strip in question must be based upon a rational utilization of the phenomena of diffusion as they affect the various elements entering into the formation of a given type of solid solution.

In fact, such is the case. One or more fundamental principles are utilized whose tendency is exclusively to modify the manner of diffusion—usually in the sense of rendering it more rapid and more nearly complete. All additional principles in the treatment are used to modify this primary action, by removal of factors which may halt or reverse its course. The criteria upon which these fundamental treatments are based can be grouped under two main heads:

1. Selection of such a chemical composition for the steel as will allow—other conditions remaining equal—the maximum velocity of diffusion for its elements in the mixed crystals.

2. Selection of a temperature at which the velocity of diffusion is at a maximum, and maintaining the steel at that heat for a sufficient time.

The second consideration is limited by the fact that a prolonged intense reheating might cause the metal to deteriorate for one or several reasons.

The first criterion (of chemical composition) may influence diffusion not so much directly—in the sense of increasing the velocity—as indirectly by decreasing the distance over which equalization must be established. In this way the conditions under which the process of diffusion must operate are modified by selecting that chemical composition in which the number of crystallization centers are maximum, and therefore the distances separating points of highest and lowest concentrations are minimum.

21. I will later have occasion to speak of the criteria of chemical composition.

Instead it is now opportune to investigate more closely the processes of the second group, which as noted have the immediate purpose to

place with the greatest velocity, other conditions being equal; and second, to keep it at that temperature for a sufficient length of time to reduce the width of the band of variant concentration to an amount which experience has shown to be sufficient to impart the desired physical properties for a particular case, or at least, to render it susceptible to acquire these properties by means of further treatments.

As is well-known, these reheatings are widely used in metallurgical practice upon raw steel (as cast) as well as upon steel which has already undergone hot work such as forging, drop forging, etc. These constitute

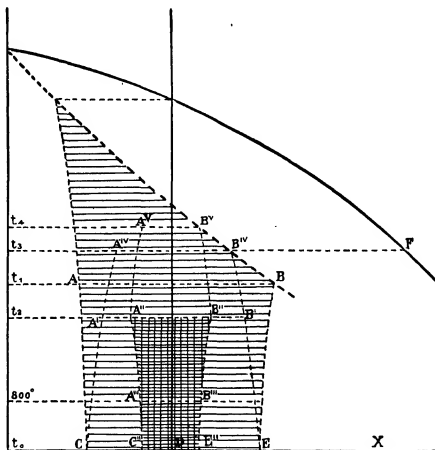


FIG. 4.—Effect of normalizing.

the foundation of the so-called preliminary heat treatments for homogeneity, or more briefly, normalizing.

We will begin as we have been doing in the preceding pages by studying the course and the effect of heating for homogeneity on a metal granule solidified around but one crystallization-germ, tracing the phenomena of diffusion of only one of the elements (which we will again call *X*) to be found in solid solution in  $\gamma$  iron.

Let us suppose that the conditions under which the crystallization of such a liquid mass of composition *D* has taken place, and that the further cooling of the solid solution down to ordinary temperature  $t_0$  has

this occurs with a velocity so small as to produce no perceptible effects even after very long periods of time. Besides this, carbon, one of the most important elements in the machine steels, cannot be found any longer as a solid solution below the temperature corresponding to the last allotropic transformation of iron, except in conditions of metastable equilibrium. This fact further disturbs the course of diffusion for this element at low temperature.

Leaving aside these considerations, suppose that immediately after being cooled to  $t_0$ , the temperature of the system be increased gradually. Diffusion of the various elements in solid solution again occurs at a velocity which makes its results felt directly upon the properties of the steel at temperatures above approximately 800°C. only. At temperatures below this limit, it takes place in noticeable measure only across intervals of space extremely small. In this interval of temperature (whose lower limit is quite variable depending upon the composition of the steel, but can be estimated for the ordinary analyses to be somewhere between 200°C. and 400°C.) the phenomena of intercrystalline diffusion only act indirectly by the segregation of metallic carbides contained in the steel.

These last-mentioned carbide segregation, occurring at temperatures below the critical interval of the steel in question, takes place in limits of space far smaller than those which are directly affected by the true processes of diffusion through a solid solution—phenomena which we are now studying. Also the two are not connected in any way in practice. The study of carbide accumulation is therefore completely extraneous to our present investigation inasmuch as it is part of the “final” or “quality” heat treatments.

**22.** Let us inquire what happens in our system when after having passed from  $t_0$  to about 800°C. its temperature continues to increase gradually.

It is clear that this first reheating has not caused the appearance of any new crystalline variety non-existent during cooling, as far as the physico-chemical properties of the solid solution in  $X$  are concerned. It also follows from the discussion immediately above, that the width of the band of heterogeneity on reheating to 800°C. will be but little smaller than the narrowest reached at the end of the first cooling. Heating at a uniform rate from 800° upward, the speed of diffusion continually increases with increasing temperature, so that the concentration, edge to center, of the crystallite will approach uniformity at an increasing rate.

After having reached a certain temperature  $t_2$ ,  $A'B'$  representing the width of the band of heterogeneity, let the mass be maintained constantly

at that temperature for a given time  $\vartheta$ . Diffusion will evidently continue, and the width of the band will decrease more rapidly the higher the temperature  $t_2$ .

Now suppose that at the end of the interval of time  $\delta$ , the difference in X-concentration at the surface and center of the crystal be reduced to the width  $A''B''$ . If the mass then cools, the phenomena of equalization will continue with appreciable though decreasing velocity until the temperature reaches about  $800^\circ\text{C}$ ., at which the width of the band will be reduced to  $A'''B'''$ , an amount yet smaller than  $A''B''$ .

Continuing the cooling under  $800^\circ\text{C}$ ., and excluding for the time being the intervention of allotropic transformation in elements, or of carbide, the width of the band does not appreciably decrease further. Hence, when the system again arrives at ordinary temperature, the heterogeneity will be reduced in value to about the value  $C''E''$ .

It may then be concluded that by heating under the conditions indicated, it is possible to reduce the width of the band of variant composition; in other words, it is thus possible to decrease the heterogeneity of the system.

23. It is clear that the increase of homogeneity obtained in a solid solution by reheating to a given point will be greater the smaller are the distances separating the loci of the extreme concentrations (maximum and minimum), and the greater is the linear velocity of diffusion of the element X.

From what has already been said, the first condition depends directly upon the dimensions of the individual crystalline elements which build about each germ of crystallization. In fact, we have seen that in each crystallite of solid solution the minimum concentration of an element is found at the nucleus, while its maximum concentration occurs at the periphery.

Thence it can be deduced, all other conditions being equal in a determined crystalline mass, that the gain in homogeneity induced by a given "reheating for diffusion" will be greater the larger has been the number of centers per unit of volume around which the crystalline elements have formed; or in other words, the greater has been the frequency of nuclei in the molten mass under observation.

As I have mentioned, I shall have occasion to deal with these phenomena further on.

In regard to the second property—the linear velocity of diffusion of a given element X in the solid solution in question—it is now opportune to observe that it depends essentially upon three conditions:

(a) Upon the nature of the metals constituting the alloy. Upon this point, about which very uncertain data exist, I will have occasion to dwell later.

(b) Upon the absolute difference in concentration of element X at

that the velocity of diffusion increases rapidly with the difference in concentration according to known laws, but this is not the moment to enter into such details. It is enough to remember that Fick's law shows that the quantity of an element diffusing in given time through unit cube of differential dimensions is proportional to the difference of its concentrations at opposite faces.

For routine application to steels we are here interested in, it is sufficient to note in regard to condition (b) that a consequence of Fick's law is the fact that a given homogeneity reheating causes effects of great intensity only during the first steps of the treatment, when the material subjected to it presents large discontinuity in concentration. When a certain point in the process of diffusion is reached, it is necessary to resort to other and much more drastic treatments to produce further noticeable improvements in properties.

(c) Upon the temperature reached and maintained during reheating. In other words, the linear velocity of diffusion due to a given difference in concentration of an element in any solid solution increases very rapidly with the temperatures at which the phenomena takes place. This is a fact absolutely general in character.

24. While experimental researches have been made with various types of nonmetallic solid solutions and for some metallic alloys (especially those of mercury) in order to establish the relation connecting velocity of diffusion and temperature, to my knowledge analogous investigations with steel have not been made of such character as to permit the deduction of general rules.

Among all the elements in steels which are present as solid solutions in iron, only carbon has been studied with anything like precision from this point of view. Attention has been directed to carbon because the velocity of diffusion at various temperatures finds immediate industrial application in the case hardening process. In cementation, the diffusion of carbides (and especially cementite,  $\text{Fe}_3\text{C}$ ) which form true solid solutions with iron, produce effects to which are added the effects of the diffusion of gaseous carburated compounds, which follow special laws known to be very much different than those followed by the others. Moreover, in the great majority of industrial carburizing, gaseous diffusion produces effects far greater than the others, totally obscuring them in the end.

These phenomena I have dwelt upon at length in another volume<sup>1</sup> and shall have occasion to touch in the following pages.

In regard to the relations connecting temperature and velocity of diffusion of the other elements—such as manganese, silicon, nickel, or chromium, which with carbon frequently enter into the composition of steel—there are no experimental data available complete enough to permit general deductions to be drawn.

<sup>1</sup> "Cementation of Iron and Steel "

To give an approximate idea of the influence of temperature on the velocity of diffusion in a metallic solid solution, I will mention briefly the results of some experimental determinations made by W. C. Roberts-Austen.

A perfectly flat surface of an alloy of lead and gold containing 5 per cent. of the latter was compressed heavily against another perfectly true surface of a block of pure lead. Maintaining the system for 30 days constantly at the temperature of 165°C., Roberts-Austen found that a perceptible quantity of gold has passed from the gold-lead alloy placed below into the surface layers of the block of pure lead placed in contact above. He found, other conditions being equal, that even a small increase in temperatures at which the two blocks were maintained, corresponded to a very large increase in the quantity of the gold passing from the block underneath to the one above.

Thus, taking as unity the velocity of diffusion of gold in lead at 100°C. under the experimental conditions mentioned above, this velocity reaches the values indicated in the second column, by increasing the temperature as indicated:

Temperature	Velocity of diffusion
100°C.	1
165°C.	From 200 to 250
200°C.	From 350 to 400
251°C.	From 1150 to 1500

Increases in the velocity of diffusion of about the same order as those above tabulated, take place in a large number of metallic solid solutions. It is also noteworthy as a nearly universal fact that the velocity increases very greatly the moment the temperature increases above a certain critical value. In the case of lead and gold this critical value lies around 150°C. It is therefore clear that a heating for homogeneity is more effective the higher the temperature at which it is performed, and naturally, the longer this temperature is maintained.

But, in practice, the attainment of the two indicated conditions designed to produce the maximum efficiency in a reheating for homogeneity is subject to the following limitations:

It is desirable to reach as high a temperature as possible, yet a brief examination of the diagram shown in Fig. 4 is sufficient to define the necessary restrictions.

In fact, if instead of arresting the increase of temperature at  $t_2$  in the first reheating cycle, as in Sec. 22 it might be possible to



that is to say, at the periphery of the individual crystallites. Surpassing this temperature even slightly, the crystalline mass will be transformed to a conglomerate of little solid grains each surrounded by a liquid film of composition  $F$ . It is true that, when the temperature of the system is maintained at a value  $t_3 + \Delta t$ , a differential amount higher than  $t_3$ , the band of variant composition applying to the mixed crystals decreases in width due to the process of diffusion, and this circumstance will in time cause the curve  $EB'B^{IV}$  to move sufficiently toward the left so that its intersection with the abscissa corresponding to the temperature  $t_3 + \Delta t$  will fall at the left of the solidus. As a consequence the system will again be totally solidified; but experience shows that in the great majority of cases in practice even the exceeding short time necessary for what may be called "re-solidification" is amply sufficient to permit certain phenomena to transpire which generally modify profoundly the physical properties of the metallic alloy under treatment. Only rarely is the damage so slight as to have an inappreciable effect.

It is not possible to describe in a general way the nature and the effects of these last phenomena. Further on, we shall see some practical examples of them. At the present time I will only mention that in many cases the damage results from processes of oxidation which take place with great rapidity in the partially molten mass under the action of the oxidising atmosphere in the furnace where heating takes place. In other cases the phenomena in question are due simply to changes in the oxygen tensions of liquid and crystalline phases in which the system is thus temporarily divided. These variations in equilibrium may at times cause the formation of new phases. Finally, sometimes these phenomena simply cause a different state of agglomeration in the non-metallic inclusions contained in the alloy under treatment.

In each case it is usually said that the metallic alloy which has suffered an alteration of the kind described has been "burned."

It is easy to see from the examination of the diagram, Fig. 4, that, other conditions being equal, the highest temperature to which an alloy consisting of homologous solid solutions can be heated without burning varies greatly with the varying degree of homogeneity of the mixed crystals constituting it, and also, as a consequence of what has already been said, with variations in the conditions of solidification and further cooling, and with the efficiency of the homogeneity reheatings which it has undergone.

Examination of Fig. 4 shows that when an alloy which has been given the heating for homogeneity described in Sec. 22, and has therefore attained a degree of uniformity represented by the line  $A''B''$ , is heated above its annealing temperature  $t_2$ , it may be taken to a considerably higher temperature before burning proceeds ( $t_3$ ) than, though the metal

significance of the solidus and border-line  $B''B^V$ . Thus, in the special case represented in Fig. 4, if—instead of continuing to increase the temperature immediately after temperature  $t_2$  has been reached with the first heating—we proceed to this increase only after  $t_2$  has been maintained for some time, it is clear that burning will begin at temperature  $t_4$ , considerably above  $t_3$ .

It is evident that preliminary heat treatments for homogeneity exercise great influence upon the high temperature a metallic alloy can safely stand without burning.

26. Even in the case in which a reheating never reaches the minimum temperature  $B^{IV}$ , at which the alloy under treatment starts to burn, the maintenance of a very high temperature for a very prolonged time may cause troubles which greatly limit the industrial application of the rules mentioned in Sec. 23, aimed at increasing the efficiency of the preliminary heat treatments to its maximum.

These troubles consist mainly in the fact that processes of agglomeration take place in the crystalline elements of all metallic alloys kept for a long time at a high temperature even if this temperature be below the solidus. The dimensions of each crystallite forming the alloy may sometimes increase in very large measure; while at the same time phenomena of "orientation" may take place within the crystalline mass. We will have occasion to see many consequences of these phenomena later, such as an increase in size of all the crystalline units in the metal, and more especially an accentuation of the effects which the direction of testing has upon the physical properties found.

Except in very special cases, the mechanical properties of an alloy which has been overheated are such that it must be considered to be in worse condition than it was before.

## PRIMARY CRYSTALLIZATION OF A TERNARY SYSTEM

27. Having outlined the manner of crystallization of binary alloys, we may now take the next step toward the more general case of multiplex systems already mentioned in Sec. 10 (Fig. 1) and which constantly occur in practice. This approach may be made by investigating a ternary alloy, which is the next more complicated than the one analyzed in the preceding pages. In other words, we will consider that portion of a ternary system rich in iron and containing two other elements, to be indicated by the symbols  $X$  and  $Y$ .

Adopting the well known method of representing such ternary system by means of a diagram in three dimensions, this amounts to the study of that corner of a cube (represented in isometric projection in Fig. 5) closely adjacent to the edge  $AB$  corresponding to pure iron.

As before, the study will be restricted to the alloys which form a homologous series of mixed crystals exclusively. There appears, then, on each of the two faces of the cube  $ABX$  and  $ABY$  a binary equilibrium diagram between iron and elements  $X$  and  $Y$  respectively such as has already been studied in detail. The planes intersect in a common ordinate,  $AB$ , representing pure iron;  $CD$  and  $CE$  respectively represent the liquidus and solidus of the system Fe:Y (iron alloyed with element  $Y$ ) and  $CF$  and  $CG$  those of the system Fe:X.

All points in the  $XY$  plane represent ternary alloys in iron, element  $X$  and element  $Y$ . Ordinates erected vertically from these various points represent the condition of those alloys with changing temperatures. Each alloy may have such an ordinate, ending at the temperature where the metal is entirely melted, and a surface formed by these ordinate-ends passes through the liquidus curves  $CD$  and  $CE$ . It, also, may be called a "liquidus surface," or more shortly, "liquidus."

In the same manner, the locus of all temperatures corresponding to the end of crystallization of the various ternary alloys, such as the point  $L$  for the alloy  $I$ , is a surface passing through the two curves  $CE$  and  $CG$ . It may be called the solidus. For the same reasons which apply in the case of the binary solidus, this last surface is also the locus of all the points representing the composition of the mixed crystals in equilibrium with residual liquids at the various temperatures.

The compositions of conjugate solid and liquid phases in equilibrium are represented by the intersection of solidus and liquidus with a horizontal plane representing the temperature.

28. Let us now take any ternary alloy, for instance that one whose composition corresponds to the vertical  $II'$ , at a temperature high enough for it to be totally melted, such as that corresponding to the ordinate of the point  $I$ . If this alloy be allowed to cool to temperature  $H$ , the intersection of the vertical  $II'$  with the liquidus surface  $CDF$ , mixed crystals will start to form around each nucleus of crystallization.

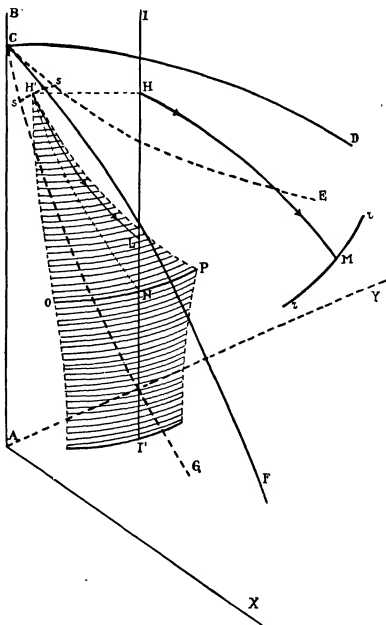


FIG. 5.—Ternary system rich in iron.

As has been done for the binary alloys, let us restrict our examination to only one crystallite. The composition of the first solid solution to separate will be represented by some point  $H'$  of the curve  $ss$  on the intersection of the solidus surface  $CGE$  with a horizontal plane passing through  $H$ . In the present state of experimental knowledge it is not

Continuing the cooling of the mass, and assuming that at every instant during the entire crystallization the diffusion of both the elements  $X$  and  $Y$  in the mixed crystals and between them and the residual liquid, takes place at such velocity as is sufficient to maintain a practically uniform concentration in the entire solid mass, it will be inferred from what we have already seen in the case of binary alloys that the point  $H'$  will follow a curve which lies on the solidus  $CGE$  and stops at the point  $L$ , when the entire mass is solidified. At the same time the point  $H$ , the composition of the residual liquid, will follow a curve  $HM$  on the liquidus  $CDF$ , and stops at some point  $M$  of the curve  $rr$ , which represents the intersection of the surface  $CDF$  with the horizontal plane passing through the point  $L$ . In regard to the position of the point  $M$  upon the curve  $rr$ , same remark applies as was made in regard to the exact position of the point  $H'$  upon the curve  $ss$ .

The discussion given in Sec. 12 covering the complete diffusion throughout the solid solution in binary systems might be repeated at this point as applying to the crystallization of a ternary system in complete equilibrium. However, as already pointed out, one may say that this case never occurs in practice. In fact, the velocity of diffusion and the time at high temperature is so limited during cooling of hot metal in industrial practice—and for that matter usually even in specially slow coolings—that the equalization in concentration between the various points of the mass is far less than that which could cause even an approximately uniform solid throughout the mushy stage.

29. This being the case, it is evident that phenomena will always appear analogous to those described for binary alloys. Let us outline the method whereby these phenomena can be represented in such a diagram as Fig. 5.

Let us still consider the ternary alloy of composition  $I$ , whose solidification begins at  $H$  by the formation of mixed crystals whose nuclear composition is represented by the point  $H'$ . When the further cooling of the mass below temperature  $H$  proceeds with such velocity as to cause incomplete equilibrium in concentration, one can consider at each instant the three following characteristic points (situated, naturally, upon the same horizontal plane) which define the maximum, average and minimum concentrations of the two elements  $X$  and  $Y$  in each crystal of solid solution.

1. The point representing maximum concentration of the two elements  $X$  and  $Y$  in each crystallite. This concentration naturally occurs at its periphery, as is the case with binary systems, and it is clear that it is represented by some point on the true solidus  $ECG$ . If  $X$  and  $Y$  diffuse at about the same speed, it will be located on a curve such as  $H'P$  lying not far from the curve  $H'L$ .

$X$  and  $Y$  in each of the crystallites. For the same reasons which have been discussed for binary alloys, it is evident that—step by step as the temperature of the system goes down—this second point follows a curve  $H'N$  which is situated totally underneath the solidus  $ECG$ , and is located between it and the axis of coordinates  $AB$ . It is also clear both by definition and from what has already been said regarding binary alloys, that this curve must meet the vertical  $II'$  at a temperature  $N$ , corresponding to the end of the crystallization of the system. This last conclusion depends upon the fact that at  $N$  the average composition of the crystalline mass is equal to the average composition of the alloy. Its validity evidently depends upon the implication that in that portion of the tri-axial diagram examined, the liquidus and the solidus do not touch each other at any other point than  $C$ , the melting point of the major constituent. This hypothesis was made in the case of the binary alloys and is now to be extended to the ternary alloys. As a matter of fact, it represents the true state of affairs always existing in the alloys which are the subject of this treatise.

3. The point representing minimum concentration of the two elements  $X$  and  $Y$  in each crystallite. It is evident that this minimum concentration is that of the nucleus.

It may again be deduced from the reasons already examined in the binary system that, as the temperature of the system lowers, this point follows a curve  $H'O$  which is situated totally below the curve  $H'N$  of average compositions, and *a fortiori* below the curve  $H'P$  of maximum concentrations. It will be also located nearer axis  $AB$  than the two preceding ones.

The third curve just defined will cut the horizontal plane passing through temperature  $N$ , at a point  $O$ , representing the concentration of the elements  $X$  and  $Y$  in the nucleus of each crystal at the instant of total solidification of the system. The same horizontal plane cuts the curve  $H'P$  in  $P$ , representing the concentration of the elements  $X$  and  $Y$  at the periphery of each crystal at the same moment.

30. From what has already been seen to happen with binary alloys, it is easy to understand that the position of the two curves  $H'N$  and  $H'O$  for an alloy of a given composition depends upon the velocity of cooling, precisely in the sense that the more rapid the cooling, the more the two curves diverge from each other and from the curve  $H'L$ . The limiting position which curve  $H'O$  approaches under drastic quenching is the vertical through  $H'$

In other words, the analysis represented by  $H'$  could be preserved at the centers of the crystallites only in case the cooling had been so sudden as to prevent the smallest accession of elements  $X$  and  $Y$  by diffusion.

Consequently, the more the mass cools, the more the three curves

coincide with the latter only in the limiting case already considered in Sec. 28, where diffusion and equilibrium is complete at all times.

Following the discussion given in Sec. 16 for binary alloys, it also follows that the temperature corresponding to complete solidification denoted by the horizontal plane *ONP* will be lower the more rapid the cooling.

**31.** From all these considerations and reasoning from analogies derived from the continuous nature of the phenomena under investigation, it may be stated that at the instant when crystallization ends, the concentrations in *X* and *Y* of the successive concentric layers formed around the nucleus of each crystal, are represented in Fig. 5 by successive points of the curved segment *ONP*.<sup>1</sup> In this segment, point *O* represents the composition of the nucleus, and *P* that of the periphery of the crystallites.

When the temperature of the system continues to decrease below the plane *ONP* corresponding to the end of the crystallization, diffusion of the two elements *X* and *Y* in the mixed crystals continues, and the two ends of the segment *O* and *P* approach the vertical *II'*.

Thus *ONP*, the "segment of variant concentration," in its successive positions and lengths assumed during cooling, describes a "surface of variant concentrations" defining the heterogeneity of the mixed crystals which gradually narrows with decreasing temperature.

The significance, behavior, and characteristics of this surface which in Fig. 5 is indicated with a curved shading, are entirely analogous to those of the band of heterogeneity studied for binary alloy. Furthermore, one can apply to the present case the exact considerations developed in Sec. 18 to Sec. 26 concerning the effects of reheating processes; their limitations on account of burning the steel; the variations in the temperature at which steel starts to "burn" caused by heat treatments, etc. Therefore it is useless to repeat here all that has already been said for binary alloys.

Nevertheless, we shall see presently in studying surfaces of variant concentrations for ternary and more complex alloys, that other phenomena specially characteristic of these more complex systems appear in addition to the phenomena analogous to those delineated for binary alloys.

<sup>1</sup> The segment *ONP* would be a straight line only in case the diffusion velocities of the two elements *X* and *Y* rigidly followed a unique interrelation during the entire

## CHAPTER V

### PRIMARY CRYSTALLIZATION OF MORE COMPLEX SYSTEMS

**32.** In the most interesting cases which appear in practice, and especially in the machine steels under consideration, it is necessary to study the transformations of systems formed by a number of constituents much larger than three. In all these cases, graphic representation of the phenomena becomes very difficult. Moreover, it can be stated that it is impossible to construct a diagram or space model for systems of more than four elements; indeed the complete graphic representation even for a quaternary system (studied especially by N. Parra-vano and G. Sirovich), is extremely complicated.

Still for purely practical purposes it is nearly always possible to reduce the study of the more complex technical problems in machine steels to the examination of relatively simple equilibrium diagrams sufficiently exact for the use intended. Such results may often be reached by first considering separately the qualitative and quantitative effects of one or two of the constituent elements as their content varies. Briefly, this may be done by selecting one or two of the elements which have the widest range in the analysis of a given type of steel, and whose effect is also the most pronounced. All the other elements composing the metal entering into solid solution with the iron, and whose analysis is substantially constant for the same type, may be lumped with the iron and considered as a "complex constituent." For instance, a plain carbon steel may be regarded, at least as far as the graphic representation of the phenomena of equilibrium is concerned, as a binary alloy. The minor constituent will be carbon, and the major constituent a "complex" containing a large amount of iron and small amounts of phosphorus, sulphur, manganese and silicon. Likewise, a nickel-carbon steel can be represented as a ternary system wherein the major constituent is the complex, and the two others are carbon and nickel.

It is easy to see why such simplified graphic representation of these systems is possible only for those which comply with certain conditions. Thus, for example, it is clear that a suppositious "complex" constituent (in the sense above mentioned) would complicate instead of simplify the tracing of the diagram, should its components give rise to distinct and separate phases within the same range of temperatures and concentrations encountered when examining the "fictitious" binary or ternary equilibrium.



restrictions upon the use of this simplified method. In particular, it is absolutely necessary to be acquainted with the various equilibrium diagrams of all the systems, binary and ternary, real and "fictitious," which can be formed with the constituent elements of the complex alloy in order that this artifice may be used with confidence. This is also true in case one wishes to examine only in a general way a hypothetical binary or ternary fictitious diagram.

But since the fiction of complex constituents will not cause any new phases to appear in the heat treatment of the soft and medium steels under consideration so long as carbon is excluded from their number, it seems timely to develop an outline showing its usefulness in the study of the greater part of the instances which happen in practice.

**33.** As just remarked, one of the components in machine steels, viz. carbon, exhibits properties which completely differentiate its behavior and functions from those of the other components, mainly for the following reasons:

(a) The linear velocity of diffusion of carbon due to a difference in concentration in the mixed crystals, is far greater than that of any of the other elements which normally join with it in the formation of the steels we are studying.

It is known that this phenomenon is essentially due to the fact that in extremely large measure the process of diffusion of carbon in solid steel is by virtue of the action of gases dissolved in the metal. At this juncture it is certainly out of place to study the mechanism of this action. It has been described in recent studies upon the cementation and case-hardening of steel.<sup>1</sup> Suffice it to say that all the experimental researches demonstrate the fact that the linear velocity of diffusion of carbon *per se* in solid solution in iron as cementite where the intervention of gases is practically excluded, is of the same magnitude as the velocity of diffusion of the other elements frequently found in steel. On the other hand, other things being equal, if diffusion of carbon in the solid steel takes place in the presence of the gases which are ordinarily dissolved in steel (notably the oxides of carbon, hydrogen, and hydrocarbons) its lineal velocity increases enormously, reaching values hundreds of times greater than those for all the other elements which with carbon comprise the analysis of the steel under examination.

(b) Other conditions being equal, variations in carbon concentration cause very much greater variations in the mechanical properties of the steel (especially under the action of certain heat treatments) than those produced by equivalent variations in concentration of most of the other elements contained in solid solution in the steel (manganese, silicon, nickel or chromium, etc.)

<sup>1</sup>This subject has been studied in my volume on "The Cementation of Iron

It is true that very small variations in absolute concentration of some of the elements contained in steel, such as sulphur and phosphorus, are sufficient to produce much greater variations in mechanical properties of the metal than those caused by even greater variations of the absolute concentration of carbon. But it is necessary to bear in mind that these elements are contained in but minute proportions in steels suitable for industrial use. Therefore, a small *absolute* variation in their concentrations is actually a great *relative* variation, and it is nothing less than these relative values which have to be taken into account, as will be clearly seen presently. It will also be necessary in some cases to give these elements special consideration, exactly as we shall soon see it is useful to do in regard to carbon.

(c) Iron carbides ( $\text{Fe}_3\text{C}$ ) in plain steel, or cementite mixed with other metallic carbides in special steels, exists as solid solution in austenite immediately after freezing is completed. Yet while the temperature is dropping from this value<sup>1</sup> to the transformation ranges of the solvent iron, they give rise to new phases by the segregation of conjugate substances thrown out of the solid solution. This is due to the presence in the austenite area of a field of incomplete solubility, called for want of a better name an "insoluble field" (Italian: *lacuna di miscibilità*; German: *Mischungslücke*; French: *lacune de miscibilité*). Such insolubility does not occur with the other constituent elements, within the limits of the concentrations occurring in machine steels.

The phenomena mentioned are noted in the equilibrium diagram by special fields whose *existence* and *significance* are essentially related to other markings in the iron:carbon diagram. Similar insoluble fields do not occur in other binary equilibria formed by the other substances taken two by two entering into the composition of any machine steels. Furthermore, the addition of these substances to the iron:carbon system may modify or obliterate some of those markings or fields representing the transformations which take place in the solid phases of such a binary system; but it can never produce new ones, provided their amounts be no greater than those in which they may appear in the steels which we are studying.

34. Therefore, from what has been said in Sec. 32, all existing conditions render it opportune to select carbon, present in the form of metallic carbides in the solid solutions, for one of the principal members of our fictitious binary diagram, and iron for the other—not pure iron, but iron containing all the other elements forming with it homogeneous molten masses or homologous mixed crystals. In the same way, the fictitious ternary diagrams to be considered in the study of some special steels shall always be made up of iron and carbon as two of the fundamental elements, and for the third the principal alloying metal in the steel under

vation. For evident reasons of simplicity, we shall begin by considering a *fictional binary* system in which the complex fundamental element is formed by the solution of but one element ( $X$ ) in iron. Then we shall see, for the alloys with which we are concerned, how possible to pass with relative facility from this simpler case to the real ones, which are a great deal more involved.

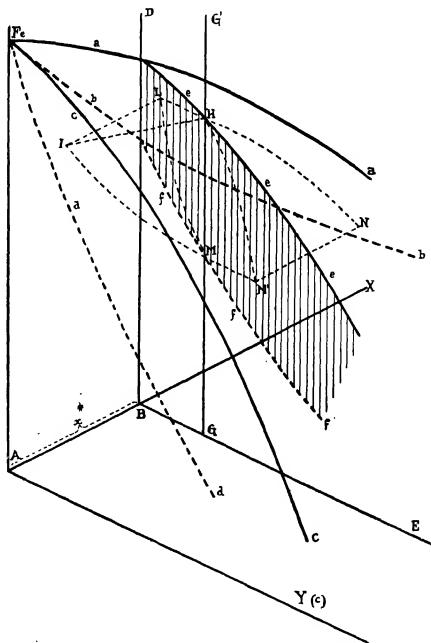


FIG. 6.—Diagram for steels containing a fixed amount of alloying element.

With this end in view, let us again examine the complete ternary diagram already described in Sec. 27 to 31 from which the binary diagram can be derived. It is formed by iron and the two elements  $X$  and  $Y$ , but

searches made up to the present time, show that the crystallization of the various binary systems composed of iron and each one of the other normal elements (except carbon) up to its maximum concentrations in ordinary specifications follows Roozeboom's first type. This comprises the systems in which only one series of mixed crystals form, and their compositions are always higher in iron than that of the residual liquid with which they are in equilibrium. Up to the limiting concentrations in machine steels these binary systems do not possess any insoluble field either in the liquid or in the solid phase.

Draw the two characteristic curves ("liquidus" and "solidus") of such a diagram between iron and another element  $Y$  upon the plane  $FeAX$  and let them be indicated respectively with  $a$  and  $b$ .

Upon the other plane  $FeAC$  draw the well known equilibrium diagram of the binary iron:carbon system, or to be more precise, of the binary system iron:iron carbide ( $Fe:Fe_3C$ ), since this corresponds to that portion of this diagram to which our deliberations will be restricted.

As has been done in the other cases, consider only the phenomena which take place at temperatures higher than the allotropic transformations of solid iron, reserving for later consideration the occurrences at or below these points. Since the highest possible carbon concentrations in plain machine steels are a great deal less than 2 per cent., which is the point at which the insoluble field between iron and iron carbide intrudes itself at the temperature of crystallization, it will be possible to restrict our present considerations to only two continuous curves, viz.: the "liquidus"  $c$  and "solidus"  $d$ .

Passing through each pair of the four curves thus traced will appear the liquidus surface  $ac$  and the solidus  $bd$  in the ternary systems. Their meaning has been amply indicated and therefore is well known.

Now consider a series of ternary alloys in which only the carbon concentration varies while the concentration of element  $X$  remains constant. Points representing the condition of these alloys at all possible temperatures will be found upon the vertical plane parallel to the  $FeAY$  plane passing through the point  $B$  representing that binary alloy of iron and  $x$  per cent. of element  $X$ .

$BD$  and  $BE$  represent the intersections of this plane with the two faces  $FeAX$  and  $FeAY$  and curves  $e$  and  $f$  the intersections of this plane with the liquidus surface  $ac$  and solidus surface  $bd$  of the same system.

Now consider all the alloys of this group as being part of a fictitious binary system whose constituents are iron carbide and the alloy  $FeX$  containing the quantity  $x$  of the original component  $X$ , and discover which of the phenomena occurring during crystallization of such a system can be represented on the plane  $DBE$ . It will be easy to see from the following analysis what actual use such a sketch derived from the true ternary diagram may have.

35. Consider any one of the alloys in the group as defined, such as the one whose composition is represented in Fig. 6 by points on the vertical  $GG'$ . Whatever may be the composition  $G$ , the initiation of crystallization will occur at the intersection of the vertical  $GG'$  with the liquidus surface  $ac$ , and, therefore, by the very definition of curve  $e$ , at the intersection of the same vertical with the curve  $e$ . Hence, it is apparent that curve  $e$  in the fictitious binary diagram retains at least a part of the significance of the ternary surface  $ac$  from which it is generated; that is to say, it locates all points where crystallization starts. It will soon be seen that it can be denoted by the name liquidus only in this restricted sense.

An analogous train of reasoning can be repeated in regard to the curve  $f$  by following the actual course of crystallization of the random alloy  $GG'$  located on plane  $DBE$ . In fact, in the limiting case where diffusion in the solid phase and between the solid and the mother liquor is able to equalize at all times the concentrations of all elements in the mixed crystals, it is clear that the end of solidification will take place at a temperature when the composition of the solid phase in equilibrium with the last residual liquid is equal to the average composition of the entire system. Therefore, in the case under examination, freezing of the alloy  $GG'$  will be complete at point  $M$ , the intersection of the vertical  $GG'$  with curve  $f$  on the solidus surface  $bd$ . This last curve in the fictitious binary diagram, thus has a meaning limited to the curve of the end of crystallization.

It is also easy to see why the two curves  $e$  and  $f$  in the fictitious binary diagram do not possess the entire meaning of the two surfaces  $ac$  and  $bd$  upon which they are located. In fact, the composition of the mixed crystals which start to separate from the molten mass at temperature  $H$  can be located on the solidus surface at  $I$ , in the manner already noted in Sec. 28. The carbon concentration in this nuclear solid solution is represented by a point  $L$ , the projection of  $I$  upon the plane  $DBE$  by a ray parallel to  $AX$ . This point  $L$  always lies above the curve  $f$  in the fictitious binary diagram, within the area between this curve and the liquidus  $e$ . This last fact is an evident result of the assumed position and curvature of the surface  $bd$ , which location is in agreement with all the known experimental results upon soft and medium steels used for machine parts.

It is therefore clear that each of the points on the curve  $f$  in the fictitious binary diagram *do not* represent the carbon concentration of the mixed crystals at equilibrium with a molten mass whose analysis corresponds to a conjugate point on the surface  $ac$  somewhere along its intersection with the same horizontal temperature plane.

As the temperature lowers step by step, the point representing the

follow a curve which starts at  $I$  and ends at  $M$ , lying entirely in the surface  $bd$ . Within this temperature interval, corresponding to the mushy stage, the point in the fictitious binary diagram representing the carbon concentration in the solid phase will follow a curved segment  $LM$ , lying entirely above curve  $f$  within the area between the two curves  $e$  and  $f$ .<sup>1</sup> In the fictitious binary diagram we may indicate this segment as the *solidus for the alloy of initial composition  $GG'$* .

At the same time, the point representing the true composition of the residual ternary liquid will follow a curved segment  $HN$  lying upon the ternary liquidus  $ac$ . This segment passes through the point  $H$  and the point  $N$  is its end as it meets the horizontal temperature-plane passing through  $M$ .

Naturally, all this discussion has value only under the hypothesis that cooling takes place slowly enough to allow the concentration of the various elements to diffuse to practical uniformity at each instant in all parts of the solid phase.

Having the positions of the two curves  $a$  and  $c$  given and assuming no discontinuity or singular points in the surface  $ac$ —it is easy to see that if  $N$  be projected normally upon plane  $DBE$  of the fictitious binary diagram, its location  $N$  will lie below the curve  $e$ . Since by definition this projection must lie in the horizontal plane passing through  $M$ , it must be located in the zone limited by the two curves  $e$  and  $f$ . It is therefore clear that, during crystallization of the system, the point representing the carbon composition of the residual liquid will follow curved segment  $HN$ , situated in the space between the curves  $e$  and  $f$  of the fictitious binary diagram.  $HN'$  (on the plane  $DBE$ ) is the trace of an element parallel to axis  $AX$  moving with curve  $HN$  as a directrix. In the fictitious binary diagram we may consider this segment as a *liquidus for the alloy of initial composition  $GG'$* .

Therefore it is seen that the curve  $e$  as well as the curve  $f$  does not possess the whole meaning of its generating surface  $ac$  in the true ternary diagram. This curve is apparently the locus of all points such as  $H$ , corresponding to the initial point of crystallization of wholly molten alloys of that average composition, but it is no longer the locus at all temperatures during the crystallization process of the composition of the mother liquor in equilibrium with the mixed crystals which have separated from it.

**36.** Once these points are established, let us see what useful conclusions can be deduced from the examination of the fictitious binary diagram which has just been described.

It is evidently convenient to rotate plane  $DBE$  of Fig. 6, which contains

<sup>1</sup> These facts could be easily demonstrated in a rigorous way, on the basis of the data we have supposed to obtain, and which really do hold for the alloys to which the present book refers. I have deemed it better not to develop these demonstrations

the projection of the diagram, upon the plane of the drawing. In Fig. 7 this is done, using the same notation as in Fig. 6.

It is clear that Fig. 7, being limited to a system of only three components (iron, carbon and X) is nothing more than a particular case of the diagram which was briefly investigated in Sec. 10 (see Fig. 1). The intervening study of the relations existing between lines and areas of Fig. 7 and the corresponding parts of the ternary diagram, from which the former are derived in a definite way, permits one to indicate precisely the real significance of the markings. Among the phenomena connected with these markings it is easy to distinguish clearly those which are

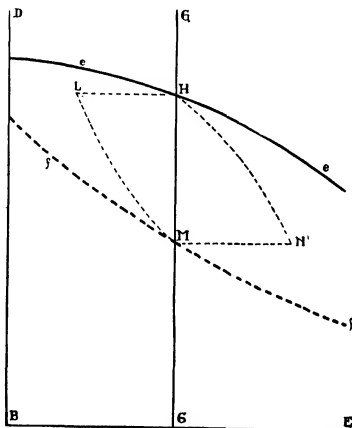


FIG. 7.—Fictitious diagram derived from ternary system.

directly due to the action of the principal constituent (in this case carbon) from those due to the influence of the other elements present in the liquid and solid solutions. The same study has established the position occupied by the liquidus  $HN'$  and the solidus  $LM$  in the fictitious binary diagram for an alloy of a determined initial composition in its relation with the liquidus and the solidus surfaces in the ternary diagram.

From these considerations it is evident that the fictitious binary diagram represents the real changes in the concentration of carbon, the fundamental element, both in the solid and in the liquid phase during the entire solidification of a given ternary alloy. On the other hand, the

sented in the same diagram, and are only reflected in it through the effects which they produce upon the carbon, effects which we shall see are essentially among their most important ones.

The practical uses of such a simplified graphic representation will be expounded immediately.

**37.** For this purpose consider the system next in increasing complexity. In other words, examine the quaternary system resulting by adding a fourth element  $Z$  to the ternary system iron : carbon : element  $X$ , which has been considered in the preceding paragraphs. Element  $Z$  is supposed to form solid solutions in iron exactly as does element  $X$  and is subject to all the conditions of equilibrium which we have assumed to exist for element  $X$  itself. It is easy to see that to represent such a quaternary system it is only necessary to construct a "fictitious ternary diagram." From this it is possible to deduce new "fictitious binary diagrams" representing the various groups of alloys of the quaternary system, in a way perfectly analogous to that which has been done for the original complete ternary system.

In detail, construct in space a true ternary diagram of the system iron : carbon : element  $Z$ . Then represent on the appropriate plane the group of all ternary alloys containing a fixed quantity of the element  $Z$ . We will thus be able to sketch, in the same way as for the system iron : carbon : element  $X$ , a fictitious binary diagram having  $X$  as the principal element and which will record the solidus and liquidus intersections for a ternary alloy with a constant proportion  $z$  of element  $Z$ .

In a similar way, let us suppose that we have constructed the fictitious diagram of the system iron : carbon : element  $Z$  in which the element  $Z$  also appears in the constant proportion  $z$ , and as a solid solution in iron forms a principal element in the fictitious binary diagram.

Now, let us take the two fictitious binary diagrams thus constructed and lay them out on the reference planes in a "fictitious ternary diagram" which will have as principal element the solution common to the two systems: that is to say, iron containing a constant amount of  $Z$  in solution. This will be located at the origin of coordinates. Along the  $X$  axis will be laid the element  $X$  containing  $z$  per cent. of the element  $Z$  in solution, and along the  $Y$  axis will be found various proportions of carbon as carbide also containing the same constant proportion of element  $Z$ .

The diagram thus constituted is represented in the usual way in Fig. 8. This can be considered in a limited sense as an equilibrium diagram comprising the alloys iron : carbon : element  $X$ , containing in addition the fixed amount  $z$  of the element  $Z$ .

In this diagram pass a plane parallel to the plane of coördinates  $FeAC$  passing through the point  $B$  on the  $X$  axis, which point corresponds to the proportion  $z$  of the element  $X$ . We will be able to trace upon this plane a fictitious binary diagram, in the same way we have already done





selected. Let us restrict consideration, for the time being, to the case in which cooling takes place slowly enough to allow diffusion to produce substantial uniformity of concentration in all parts of each of the phases.

Curves  $e$  and  $f$  are located at the intersections of plane  $DBE$  with surfaces  $ac$  and  $bd$ . All points of these curves correspond respectively to the start and end of crystallization of the alloys whose average composition is represented by the projection of the points in question on the  $XY$  plane.

One can repeat nearly the same remarks made in Sec. 35 for the preceding case, regarding the compositions of the solids in equilibrium with the successive residual liquids as the crystallization gradually proceeds. The only essential difference between the two cases (which may be distinguished as the "true" ternary system and the "fictitious" ternary system respectively) is entirely similar to the difference existing between the "true" and "fictitious" binary diagrams. It is this: the points representing the composition of the solid phases in equilibrium with mother liquor throughout the mushy stage do not lie any longer upon the surface  $bd$  representing the end of solidification.

It can be demonstrated that these points (similar to point  $I$ , at the start of crystallization) are located upon a segment of curve  $IM$  lying entirely above the surface  $bd$  but intersecting it in the point  $M$ , corresponding to the end of crystallization of the alloy considered.

Without entering into a rigorous demonstration—necessarily long yet superfluous here—it is easy to understand why this may be true. First bear in mind that the composition of the solid phase in the fictitious binary system represented upon the plane  $FeAC$  has been demonstrated to be located on a curve located above curve  $d$ . Plane  $FeAC$  corresponds to  $x = 0$ . Evidently also the locus of the solid phase would be located above surface  $db$  in systems directly derived from  $x = 0$  simply by adding sufficiently small quantities of the element  $Y$ . Now, if any point such as  $O$  on the segment  $IM$  be supposed for the moment to be located below the surface  $bd$ , such supposition would lead to consequences openly at variance with the definitions and with the known fundamental properties of the diagram. In fact, under this hypothesis, the vertical ordinate passing through  $O$  would cut the surface  $bd$  in a point  $W$  corresponding to a temperature *higher* than that corresponding to point  $O$ . But it is known that the point  $W$  corresponds to the end of crystallization of the alloy of average composition  $OW$ . It therefore represents the *lowest* temperature at which mixed crystals of uniform composition  $OW$  may be in equilibrium with the last traces of a mother liquor whose composition is located at some point on the intersection between the surface  $ac$  and the horizontal temperature plane passing through point  $W$ . But, by the very definition of segment  $IM$  mixed crystals of the same composition

rium with a residual liquid (corresponding to some point on the surface  $ac$ ). This can occur only when  $O$  is at or above  $W$ , and it is evidently not possible that the point  $O$  can be placed underneath as originally supposed. It is indeed true that the proportion of the element  $Z$  in the mixed crystals  $O$  and  $W$  is not the same; but we positively know that in the crystals located at  $W$  this proportion has by definition the value  $z$ , while in the crystals  $O$  it is certainly less than  $z$ , which it reaches as a matter of fact only when solidification is complete and the solid solution possesses the same composition as that represented for the alloy  $GG'$ . This difference of the concentrations of the element  $Z$  adds therefore additional weight to the deduction made regarding the position of curve  $IM$ .

At any rate, for reasons entirely similar to those mentioned when considering a true ternary diagram and depending upon the curvature of the surface  $bd$ , it is clear that the projection  $LM$  of the segment  $IM$  upon plane  $DBE$  will lie totally above curve  $f$  within the area between this last curve and curve  $e$ .

Similar considerations, which would be superfluous to repeat, will locate the pseudo-liquidus curves  $HN$  and  $HN'$  in a path conjugate to the one indicated for curves  $IM$  and  $LM$  respectively (see Sec. 35).

39. Therefore, even in the more complex case now examined, one may construct a "fictitious binary diagram" entirely like the one which was obtained in the simpler preceding case (Fig. 7, and Sec. 36). The same considerations made at that time can be repeated.

Without entering into useless details or repeating similar reasons it follows that analogous conclusions may also be reached for more complex alloys containing three, four or even more elements, each one in a constant percentage, together with the two fundamental elements iron and carbon present in variable proportions. Naturally this remark is true only when the fundamental conditions as to solubility and continuity are taken as a necessary and essential basis. But as has repeatedly been said, these always obtain in practice for the steels belonging to the group to which we are limiting our study.

It is now timely to remark that the preceding discussion and its accompanying sketches have visualized an idea cleverly enunciated and verified by Guillet.<sup>1</sup> He postulates the existence of "equivalent values" for the various components entering into metallic alloys, especially as regards the solid solutions and their decomposition products. In our graphic representation, these "equivalents" would be defined as the quantities of each of the half dozen or more component elements which cause

equal displacements of the pseudo-solidus and -liquidus in a fictitious binary diagram for a series of alloys of given composition in the two principal elements selected.

The experimental data available in regard to this subject at the present time do not permit the formulation of general conclusions, nor the quantitative application of the graphic method to the complete and precise study of a practical case. Nevertheless we are going to see later how this representation may find convenient application in the partial study of some phenomena which take place in our steels. It would certainly be of a great industrial importance to gather a collection of experimental data, so as to define better the conceptions introduced by Guillet, and to coordinate his conclusions in a series of complete and well defined equilibrium diagrams. To reach this point, however, it will be necessary to gather a very large amount of experimental data.

40. Using the conclusions already attained for the more complex cases as a basis it is possible to represent graphically the phenomena which take place when the cooling velocity of the system is high, too high to permit diffusion to proceed in such measure as to equalize the concentrations of the various constituents at each instant in each phase.

To do this, take as starting point the aggregate data which has been established in Sec. 29, 30 and 31 in regard to the true ternary system. A brief survey of this information in relation to the elements of the fictitious binary and ternary diagrams defined in the preceding pages shows instantly that all the considerations already made in the simpler cases may unquestionably be applied to the more complex cases represented in the corresponding diagrams.

In fact we have seen that the fictitious diagrams of even the very complex systems always show in a well defined manner the segments of curves locating the points of initial and final crystallization for every alloy of a determined average initial composition. Moreover, in the same diagrams, the curves or surfaces denoting the locus of the start and the end of crystallization also bound the space within which must lie all the curved segments which determine the variation in composition throughout the mushy stage.

Being merely an exact repetition of the same reasoning already detailed, I do not deem it opportune to again go over the minute analysis of the familiar processes connected with the incomplete development of diffusion, nor to a detailed explanation of the graphic representation of the phenomena.

I believe it is enough to mention the fact that the fictitious diagrams of the more complex systems will also be able to show the "curve of average composition" of the mixed crystals and the "curve of minimum concentrations" of the principal elements in solid solution, together with

talline element. These curves have a course entirely similar to those described in the simpler cases, and bound an area of variant concentration whose significance is also like that analyzed for the preceding cases, when the former is interpreted according to the well defined values which have been assigned to the lines and fields of the fictitious diagram.

Thus, diagrams may always be constructed entirely similar to those which were sketched when studying the binary and ternary systems (see Fig. 1, 2, 3, 4 and 5). Disregarding the difference in composition of the various phases inherent to the selection of different fundamental elements, simple and complex, the only difference in form which is found between the points, curves and surfaces of a true diagram of a simple system and those of a fictitious diagram of a complex system, lies in the fact that in the latter both liquidus and solidus vary upon changing the initial composition of the alloy, and do not coincide any longer with the curves or surfaces representing the start and the end of crystallization.

41. Regarding the steels to which this study applies, we have already seen in Sec. 33 that for various reasons it is convenient to represent the various systems—always very complex—by means of fictitious diagrams belonging to two fundamental types:

(a) A fictitious binary diagram between carbon and iron, the iron containing a certain amount of other chemical elements in solid solution. This diagram lends itself well to the study of the so-called "carbon steels," in which none of the elements to be lumped together with the iron will give the steel any properties which notably overwhelm those produced by any of the others.

(b) A fictitious ternary diagram between iron, carbon and another metal, iron again carrying the many elements in solid solution. This diagram lends itself well to the study of those so-called "special steels" in which one of the constituent elements other than carbon has a preponderant influence upon the properties of the steel, both because it is contained in the steel in greater proportions and because its specific effect is more intense than those of the other components, even at parity of concentration.

Sometimes it may be convenient when studying some practical cases to take into consideration another method of graphic representation of the more complex systems, founded directly upon the application of Guillet's conception of equivalence already mentioned in Sec. 39. This method uses a three-dimension diagram which may be defined in the following way:

(c) A diagram having three rectangular axes. The vertical axis represents pure iron, and distances up from the origin represents temperatures. Upon one horizontal axis is taken the concentrations of the second principal element, which in our cases we have seen normally is

thetical equivalent of the other elements contained in the alloy taken all together.

It is clear that a complete model thus constructed could lend itself to an exact graphic examination of the manner of crystallization of the various alloys and of the various phenomena of transformation which take place in them, only in case we knew with precision from complete experimental data the "concentration equivalents" of each one of the elements whose concentrations must be lumped together on the third axis of the diagram. In fact, it is clear that such detailed knowledge is necessary if we simply want to locate an alloy of given composition upon the diagram. This holds true even when the average value corresponds simply to the sum of the concentrations of each element.

However, inasmuch as these equivalents have not been determined, nor is it known if an absolute value may be attributed to them with confidence, it is clear that diagrams constructed according to Guillet's ideas may not be used in any of the cases in which it is necessary to establish any quantitative datum.

Such an approximate diagram may present some interest when one merely wants to represent graphically and but qualitatively the phenomena taking place in a given complex alloy.

It is clear that a diagram of type (c) does not differ in construction from the *true* ternary diagram, which we have already studied (see Fig. 5 and 6). In fact, we find again in it all the characteristic lines and fields already analyzed in Sec. 27 to 31. Therefore it is not necessary to repeat its detailed examination, which anybody can easily accomplish on the basis of Fig. 5 and 6, by substituting the sum total of concentration "equivalents" for one of the principle elements such as *Y* in Fig. 5.

The only difference between the two cases consists in the fact that points and lines in diagrams of the third type (c) do not represent any definite analysis among the family of alloys except in so far as the percentage of iron and carbon are concerned. On the other hand, in the *true* diagram of a ternary system, as well as in the "fictitious" diagram of more complex systems constructed as specified just above in (a) and (b) the significance of each line, point and area representing a phase or having other meaning is always perfectly correlated to the initial composition of each alloy.

In the special cases to which this present study is limited, it is almost always of advantage to use one of the first two types of simplified diagram—either (a) or (b).



## PART II

### EFFECTS OF DIFFUSION UPON SECONDARY CRYSTALLIZATION

#### CHAPTER VI

#### GENERAL REMARKS UPON THE RELATIONS BETWEEN DIFFUSION AND SECONDARY CRYSTALLIZATION

42. Let us now outline the phenomena which take place in a completely solidified machine steel when its temperature falls below the value corresponding to an allotropic transformation of one of the constituents.

In our steels, the only allotropic transformations which must be taken into consideration are those of iron. Only for special steels very high in elements other than iron (for instance high nickel steels) it may become necessary to take into account the inversions of some of those elements, but these can not be classed as soft or medium steel for the construction of machine parts.

Having already noted the inherent function of iron, in that its concentration is always higher than that of all the other elements forming these steels, we may restrict our considerations to the simpler case of the allotropic transformation of the element which acts as "solvent" in the system under investigation.

It is known that practically pure iron in cooling below the temperature of its solidification (about 1520°) undergoes three reversible transformations, revealed by three arrests at 1410°, 910° and 790°C.

Little data is available upon the first transformation point, corresponding to the passage of delta to gamma iron. It has not yet been possible to isolate delta iron, and its properties are therefore not known. In carbon steels it seems that this transformation ceases to take place when the carbon concentration reaches or exceeds about 0.3 per cent. If a curve could be drawn in the iron carbon equilibrium diagram representing the beginning of the  $\delta \rightarrow \gamma$  transformation, it would intersect the solidus at this concentration. It is very probable that precise and complete knowledge about this first transformation of iron will throw great light upon various occurrences frequently observed in industrial treatment of steel, and which have not been explainable by means of the lines and areas in the equilibrium diagram based only upon the other transformation-points. At any rate, the data available upon this subject are too scarce and uncertain to be taken into the slightest account in a study of



The nature of the second reversible transformation—corresponding to the passage of gamma into beta iron—has been the object of numerous researches, which, however, have not yet totally clarified the problem. Even supposing the reader to be acquainted with the recent literature upon this point, I must point out that I cannot dwell upon its nature, since prominent experimentors have even doubted the existence of such an allotropic transformation, basing their train of reasoning on the self same researches which others have made to establish its nature. In fact, the results of the experimental researches now available are essentially negative in character, yet certain facts have been discovered which are difficult to harmonize with the older accepted iron:carbon diagram showing a definite phase called beta iron, yet it is not yet possible to substitute new lines and areas to take its place.

Nor would I need to modify or substantially extend my conclusions should I include in this study what might be called the positive information now available regarding beta iron. This because as has been remarked often before I intend to limit consideration to those phenomena strictly dependent upon the chemical composition of the various parts of the crystalline elements of the steel; mentioning only in passing a brief outline of those facts which depend essentially upon the morphology of the metallic constituents—that is to say, upon their crystallography, using the term in its limited and precise significance.

Various experimentors have reached much more concordant ideas (based upon a very large mass of experimental material) upon the nature of the third reversible transformation of iron, corresponding to the passage of beta iron into the alpha modification or from gamma direct into alpha iron.

**43.** The rearrangement of phases accompanying allotropic transformations of iron take place according to laws entirely similar to those governing crystallization of the same alloys.

This statement holds in the binary systems iron:carbon, iron:manganese or iron with any of the other elements entering into the composition of machine steels, as well as in the more complex alloys of the same elements within the limits of concentration to which I intend to confine my study. In fact, the transformation ranges are zones wherein new solid solutions are forming. The concentration of the various elements existing in saturated solid solution with low-temperature modifications of iron as a solvent is always less at the temperature of equilibrium than the concentration of the same elements in saturated solid solution with another modification which is stable at the higher temperature. Evidently this statement holds for the temperatures at which both types of solutions, that is to say, both phases, are reciprocally in equilibrium.

From this it follows that we may base our study of the transformations

ciples already developed in the study of the first transformation (crystallization) of a liquid solution of the same composition. This under one condition however: that the original solid solution under consideration has been maintained at high temperatures during a sufficiently long period of time to cause the concentrations of its various components to be practically uniform.

When the last condition does not hold, it is clear that we will have to add to the reasoning developed in Part I the necessity of taking under examination a band of variant concentration in the region which represents the phase stable at temperatures above the transformation point. Under absolute homogeneity, of course, such a band contracts into a vertical ordinate.

I will analyze the two cases separately.

44. Beginning with the first and simpler case, in which the solid solution stable above the transformation is exactly homogeneous, the application of the reasoning utilized when discussing crystallization will deduce some new conclusions. These are due, for all elements except carbon, to certain differences of a purely quantitative nature in the way in which the different phenomena occur. In the case of carbon, a large proportion of the steels we will consider have an analysis which locates them in a region of insolubility in the equilibrium diagram extending downward from the eutectoid horizontal.

Experimental data known at present are very scarce and uncertain concerning the phenomena connected with the alloying elements other than carbon. The only fact that can now be affirmed with apparent certainty is that the transformation range<sup>1</sup> is very small. At least this remark applies in case the transformations take place without hysteresis of noteworthy amount.

The consequences of a narrow transformation range are easily inferred from the reasoning already applied to the process of solidification expounded in the preceding pages and may be summarized in the following remarks, which apply to the various elements usually present in machine steels, *carbon excepted*.

(a) If as an effect of suitable diffusion, the homogeneity of the phase stable above the transformation is practically complete, so that the composition of its individual crystallites (and of the whole mass) may be represented on the equilibrium diagrams by points, then the composition of each crystallite at temperatures below transformation can be repre-

<sup>1</sup> That is to say, the temperature at which the allotropy of iron first commences to make itself felt, either by evolution of heat, change in magnetism or some other physical property, is comparatively close to the temperature when these changes are at maximum intensities and to the temperature where the effects die out. Such

sented by a band of variant concentration so narrow that in practice these strips can be considered as lines.

(b) When the composition in the crystallites, edge to center, is appreciably non-uniform in the high-temperature phase, the band of heterogeneity induced in this phase by the special treatments which it has undergone, is not very much modified after transformation to the phase stable at lower temperatures.

Thermal hysteresis is frequently observed and is sometimes very intense in the transformations with which we are now concerned. It may sometimes substantially modify the course of the phenomena indicated above. But our actual knowledge in regard to the special effects of hysteresis is too fragmentary to be taken into account in practice. However, it is very probable that a precise study of this problem might illuminate the nature of many occurrences now known and be utilized in steel making and heat treating practice without being able to place them rationally in the outline diagrams of binary equilibria as experimentally determined.

It is not possible, therefore, to present a complete graphical study of the group of phenomena indicated in the last paragraph. It is yet a blank page to be written in the future complete treatise on heat treatment.

45. Our knowledge regarding carbon is less incomplete on such points.

As has been seen in the case of solidification of liquid metal into crystalline solid, it was unnecessary to take into account the presence of regions of insolubility when examining the actions due to any one of the elements occurring in machine steels up to the maximum concentrations there found.

As a consequence, the partial or complete systems of any of these elements with iron (carbon included) could be represented during solidification by means of schematic diagrams perfectly similar to each other. On the contrary, in the case of the transformations due to allotropy in iron, areas appear in the iron-carbon equilibrium diagrams which denote a limited solubility or complete insolubility. Such insoluble areas, so-called, may be again found although substantially modified in the majority of the other more complex systems.

Supposing the reader to be acquainted with the latest information about the equilibrium phenomena proceeding in an iron : carbon system possessing perfect homogeneity of concentration in all the mixed crystals of each series, I shall refer my remarks to that part of the familiar diagram which contains the low-carbon concentrations we are interested

it includes cast irons as well as steels. As already remarked, it is not necessary to stop and explain the meaning of the various elements of Fig. 9. It is reproduced only in order to establish clearly which one of the proposed variations I intend to discuss. Perhaps it is superfluous

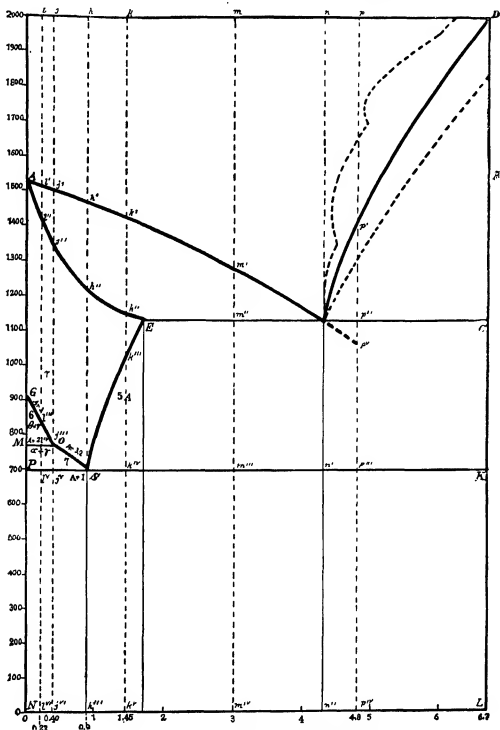


FIG. 9.—Iron:carbon equilibrium diagram.

to note that, inasmuch as we are concerned only with steels with low

Let us start, therefore, by investigating briefly the phenomena which take place during the more or less rapid cooling to room temperature of binary iron : carbon alloys ranging from zero to about 0.9 per cent. in carbon.

In Fig. 10 is reproduced an enlarged portion of that part of the diagram containing carbon concentrations up to the eutectoid percentage, and covering the entire interval of temperature within which takes place all transformations of the three allotropic forms of iron—alpha, beta and gamma. It does not take into account some of the hypotheses recently proposed, such as Benedick's "ferronite," because the required experimental evidence certainly is far from complete.

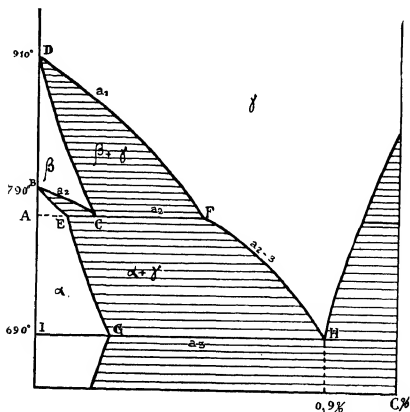


FIG. 10.—Transformation range in hypoeutectoid steels.

Excluding the possibility of any heat treatment liable to materially interfere with or arrest the transformation of beta into alpha iron, such as a drastic quenching from a temperature between A and 790°C.; and considering only such slow cooling as to permit the complete allotropic transformations, it is easy to see that the effects of the two transitions which take place in the areas *BEC* and *DCF* simply accumulate upon each other. As a consequence, one may greatly simplify the considerations of general qualitative character without materially affecting the soundness of the conclusions, by merging the two above mentioned areas

(Section 46) in which are represented only the effects of the complete transformation. It would be easy to demonstrate that the conclusions reached for this case are entirely valid for the complete diagram of Fig. 10. However the fundamental ideas upon which such demonstrations should be founded are so obvious that it is not necessary to linger upon them. To establish them it is simply sufficient to compare the method of crystallization as sketched in the two diagrams.

Therefore, the simplified diagram Fig. 11 is to be used. In order that the diagram may be more legible, the distance  $IG$  has been increased much more than its correct length as determined by experiment. Thus, the width of that portion  $GH$  of the area of insolubility existing between alpha iron and iron carbide has been narrowed so that in the figure the concentration of carbon in saturated alpha iron appears a great deal larger than its actual value. It will be more clearly seen further on that this modification evidently cannot alter the value of a qualitative discussion in any way.

## CHAPTER VII

### THE BEGINNING OF SECONDARY CRYSTALLIZATION IN HOMOGENEOUS AUSTENITE

46. Let us, therefore, investigate the behavior of an iron : carbon alloy of composition  $L'$  (Fig. 11) at a temperature  $L$  at which the iron is totally in the gamma form. Let us suppose that this alloy has been maintained at a sufficiently high temperature during such a lapse of time as to allow the carbon concentration to become uniform in all points of the metallic mass.

If we allow such an alloy to cool more or less slowly, the course of the phenomena which take place at temperatures above  $IGH$  ( $690^{\circ}\text{C.}$ ) will be entirely similar to that already studied for the liquid : crystalline transformation. Attention will also be restricted for the moment to the space occupied by a small crystalline mass forming around a single germ of transformed alpha iron. For this elementary solid we will also obtain in the diagram first, a curve of minimum concentrations of carbon in the alpha solid solution (curve  $d$ ); second, a curve of the average composition of the alpha phase which has formed at each temperature (curve  $c$ ); and third, curve  $b$  of maximum concentrations which coincides with what physical chemists call the "solidus of the transformation  $\gamma \rightarrow \alpha$ ."

In regard to these three curves we may repeat the same observations already made in Sec. 16 to 19. The only difference between the two cases is quantitative. The velocity of diffusion of carbon in the alpha solid solution, as well as in gamma solid solution or austenite, at the temperatures at which the  $\gamma \rightarrow \alpha$  transformation takes place, is a great deal less than that existing at temperatures near melting. Consequently, given equal dimensions of the crystallites, it is necessary to have much slower coolings at temperatures below  $900^{\circ}\text{C.}$ , and much more drastic reheatings and quenchings in order to obtain substantial displacements in the location of curves  $c$  and  $d$ .

However, it should be noted that the linear velocity of diffusion is not the only factor which must be reckoned with in any treatment to produce homogeneity in a steel. For instance, given quite small crystals of ferrite and cementite, equalization may be more rapid even at temperatures just below the transformation than at a considerably higher temperature, which produces coarse austenitic grains. This is for the reason that the distance between the points corresponding to the extreme concentrations of carbon in each crystalline element or colony is a great

composition of the alloy and velocity of the cooling. These can be divided in two groups based upon the total composition of the alloys considered.

The first group comprises those cases in which the line representing the average composition of the alloy is represented by a concentration-vertical

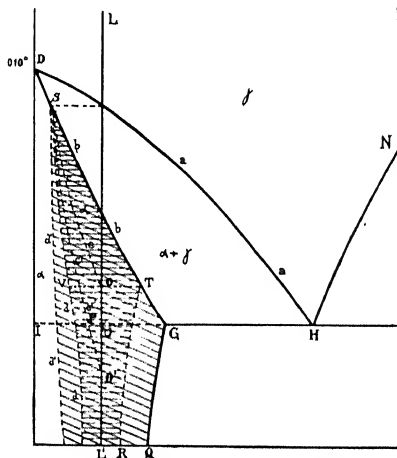


FIG. 11. Precipitation of carbon from ferrite.

such as  $LL'$  which intersects  $IGH$  at the left of the point  $G$ , corresponding to the carbon content of saturated mixed crystals of alpha iron at  $690^\circ$ .

In this case the manner in which the transformations occur and the final result vary a great deal, not only quantitatively but qualitatively, with the variation of the cooling velocity.

In fact, if the temperature of the real end of the transformation—corresponding to  $O$ , the intersection of curve  $c$  with the vertical  $LL'$ —is higher than the eutectoid temperature  $IGH$ , we again have a system entirely similar to those which have been studied under primary crystallization. When the transformation is complete, each transformation germ will be surrounded by a structural element comprised of solid solutions of



The only difference between the present case and the one studied in Part I lies in the fact that the curve  $TR$ , which is the right boundary of the band of variant concentration, may meet the curve  $GQ$ , which is the left boundary of the region of insolubility of carbide in alpha iron, and thus require the formation of a new phase. However, with our steels, this intersection does not occur in practice, since the two curves  $TR$  and  $GQ$  approach verticals passing through  $T$  and  $G$ ; or if it does occur, it is at temperatures so low as to prevent the theoretical transformations to go forward appreciably.

47. It is clear that the position of the point  $O$  depends upon the initial composition of the alloy  $LL'$  and the velocity of cooling (that is to say, upon the position of curve  $c$ ).

If for a given cooling velocity the initial concentration of carbon is high enough (although remaining below the value  $G$ ), or else if for a given initial carbon concentration the cooling velocity is high enough to keep curve  $c$  diverging considerably toward the left of the solidus  $b$ ,<sup>1</sup> it will be possible for the point  $O$  to fall below the eutectoid temperature  $IGH$ .

Let us suppose, for instance, that the same alloy  $LL'$  be cooled more rapidly than before so that the curve of average composition of the  $\alpha$  phase assumes the position  $c'$  to the left of the preceding position  $c$ . In this case the temperature of the end of the transformation will be the one corresponding to the point  $O'$ ; in other words below the eutectoid temperature  $IGH$ .

When this occurs it is evident that at the eutectoid temperature ( $690^{\circ}\text{C}.$ ), the periphery of the structural element under examination will be constituted of saturated solid solution in alpha iron of composition  $G$ , in equilibrium with a residuum of solid solution in gamma iron of composition  $H$ . A further subtraction of heat from the metallic mass will cause this residual gamma iron to be transformed at constant temperature or isothermally into the eutectoid of the same total composition ( $H$ ). Such decomposition results in an intimate mixture of small crystals of two constituents stable at lower temperatures. One phase consists of particles of composition  $G$  and the other of particles of composition represented by the right hand end of the eutectoid horizontal. This horizontal represents the top width of a band of insolubility of iron carbide in alpha iron, or a region of heterogeneous equilibrium. It is known that the right hand border is the ordinate representing the composition of cementite (iron carbide, or  $\text{Fe}_3\text{C}$ ) or a line near by representing a solid solution of alpha iron in cementite.

It is useful to keep in mind the fact that the composition  $H$  of the

<sup>1</sup> See Sec. 19.

eutectoid thus formed is practically independent not only of the initial composition of the alloy, but also of the cooling velocity. The significance of points  $P$ ,  $U$  and  $H$  need not again be defined. Given their evident meaning, it may be easily demonstrated as has been done before in Sec. 13 that the following relation holds when the transformation is completed.

$$\frac{M_p}{M_e} = \frac{\overline{UH}}{\overline{PU}} \quad (1)$$

where  $M_p$  is the mass of alpha mixed crystals of average composition  $P$ , and  $M_e$  is the mass of eutectoid of composition  $H$ .

From this equation it is seen that the proportion of eutectoid will be greater the more the curve  $c'$  is shifted toward the left; namely, the more rapid has been the cooling.

From the preceding remarks it also follows that it is always possible to select such a cooling velocity as to prevent the appearance of eutectoid in alloys of carbon content less than  $IG$ . Other rates of cooling will cause the appearance of a proportional quantity of eutectoid whose ratio may vary from zero to a maximum equal to  $\overline{SL''} \div \overline{UH}$ .

If, on the other hand, we start from an alloy in which the initial concentration of the carbon be *higher* than that of the saturated mixed crystals  $G$ , it is clear that material of eutectoid composition will always be present. Each variation in cooling rate will cause the ratio between the quantity of alpha mixed crystals and the quantity of eutectoid to vary; but it can never be slow enough to obliterate the pearlite completely.

The preceding argument neglected the fact that the width of the insoluble band or the field of heterogeneous equilibrium is subject to change as the temperature drops below the eutectoid temperature  $IGH$ . These variations are partially represented by the inclination of the curve  $GQ$  away from the vertical but in reality are of a size very much smaller than any phenomena which cause changes in the characteristics of the alloy noticeable in practice. Besides this, the variations in solubility of cementite in alpha iron occur at temperatures where the velocity of diffusion is so small as to render their effects practically negligible in the great majority of cases.

48. The reasoning outlined in the preceding sections, has been based purely upon a simple extension of the conclusions which were obtained when studying the formation of a solid solution from a molten mass. Now it is true that such an extension is wholly justified, within the approximate limits which are of interest in heat treating practice, when the train of thought is restricted to the *nature* of the constituents appearing in the transformed alloy. But it is no longer even approximately justified, when we wish to reach conclusions concerning the forms and the positions assumed by the phases produced by the decomposition of

austenite, nor the relation between the quantities of these structural elements.

This is due essentially to two series of phenomena, whose origins lie in entirely similar situations and whose consequences reinforce one another. Therefore they cannot be studied separately.

We know from the very definition of equilibrium that all the conclusions reached when studying the separation of a solid solution from a molten mass are valid only in case diffusion is rapid and substantially equalizes all heterogeneity in each phase. In practice, this takes place with a sufficient approximation to reality that the hypothesis may form a working basis. In other words, processes of diffusion due to differences in concentration do take place in the mother liquor during all that interval of temperature within which crystallization *really* occurs, with velocity sufficient so that the concentrations of the various components of the liquid phase may be considered as practically uniform at each instant in all the points of that phase itself.

On the other hand, we also know from a somewhat detailed study that such an ideal homogeneity never occurs in solid solutions. In fact it has been seen how it is always necessary to draw a band of variant composition down through the field representing mixed crystals to visualize the amount of residual heterogeneity of this phase at various temperatures and after various heat treatments.

Therefore, in the study of new phases resulting from allotropic transformations in iron, it is not only necessary to take into account the very slow velocity of diffusion at these relatively low temperatures, but also the additional heterogeneity inherited from the primary solidification and uneffaced in the subsequent history of the metal.

49. With this in mind, I will restrict my remarks for the time being to the examination of the effects of retarded diffusion upon the forms and positions of the various constituents born in the solid iron-carbon alloy. This alloy, however, will be thought of as having undergone that heat treatment, very rare in practice, which will thoroughly equalize the concentrations of the various components in solid solution in gamma iron. It is evident that the composition of such austenite, stable at the higher temperatures, may be represented by a point which traces a vertical line in the diagram as the temperature of the system lowers.

Should completely homogeneous austenite occur in practice the identical considerations already developed when discussing the crystallization process could be repeated for it, as far as the *beginning* of the  $\gamma \rightleftharpoons \alpha$  allotropic transformation is concerned. Again, during a slow cooling, the transformation point would be reached simultaneously in all parts of the mass. Separation of the phase stable at the lower temperatures

same temperature would start around centers of crystallization whose distribution depends directly upon the nature of the decomposing solid solution and upon the cooling conditions of the metallic mass.

Let us suppose for the moment that the secondary transformation might in reality start exactly in the way just indicated, and let us examine briefly how it should proceed further.

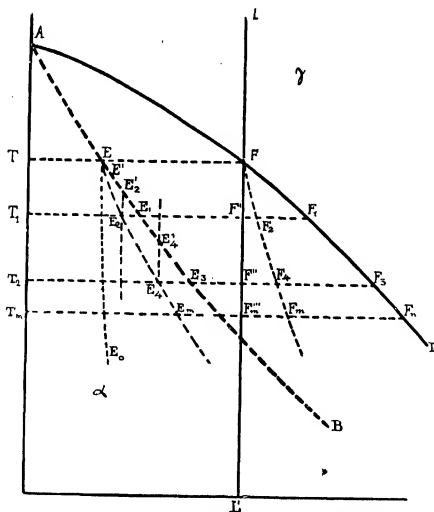


FIG. 12.—Early stages in the decomposition of a particle of homogeneous austenite.

When the iron: carbon alloy of composition  $LL'$  (Fig. 12) during uniform cooling has reached the temperature  $T$  of incipient transformation, mixed crystals will begin to form around certain nuclei. After a very short time  $t$  the temperature will drop further through the small interval<sup>1</sup>  $T - T_1$ .

<sup>1</sup> In Fig. 12 the interval  $TT_1$  has been taken very large as compared with the total interval of crystallization, with the aim of making the diagram clearer. The same remark may be repeated for the following interval  $T_1T_2$ . It is evident that this distortion or magnification does not in any way vitiate the accuracy of the conclusions I shall develop further on. It is equally evident that it would not have been practicable to represent minute distances to exact scale. To do so, I should have taken the partial intervals so small as to render them practically invisible.

Upon reaching the temperature  $T_1$ , small masses of alpha solid solution will have formed at those crystallization centers. In these minute masses the carbon concentration will have an average value  $E_2$ , a value lying between  $E$  and  $E_1$ . Its composition will vary from the center to the periphery from a value corresponding to a point  $E'$  (situated to the right of  $E$ , but very near to it) to the value  $E_1$ . At the same time the carbon concentration in the neighboring regions around each nucleus increases till it reaches the value  $F_1$ , corresponding to equilibrium between austenite and the alpha crystals at temperature  $T_1$ . I have already discussed the reasons and the mechanism of this action in Part I.

50. If, now, we endeavor to determine the quantity of alpha mixed crystals which have separated from the gamma solid solution at temperature  $T_1$ , we will see that it depends directly upon the diffusion velocity of carbon in austenite at that temperature.

It has been postulated that diffusion is slow, a supposition which corresponds precisely to reality in the case now under study. Hence homogeneity of carbon concentration throughout the mass of austenite is impossible of attainment in the intervals of time used in any possible cooling, however slow. It is clear then, that at the end of the minute time interval  $t$ , the residual austenite will be composed of two distinct parts; as follows.

(a) First, a heterogeneous region immediately surrounding the germs of alpha iron first to appear. In these regions the carbon concentration will vary from a maximum value  $F_1$  for the shell in immediate contact with the alpha solid solution, to a minimum value  $F$  corresponding to the external surfaces of the region which has been disturbed by diffusion. It is clear that the distance of that surface from the alpha crystals, and therefore, the radius of the volume in which the carbon concentration is higher than  $F$  will be greater, the higher the velocity of carbon diffusion, other things being equal. In other words, given a certain velocity of diffusion, the volume of the heterogeneous portion will be the greater the longer the time  $t$  during which the temperature  $T$  falls to the temperature  $T_1$ .

(b) A second part, formed by all the remaining mass of the gamma mixed crystals interposed between the regions defined in (a). In this portion in which the effects of the carbon accumulation in the central zones have not been felt, the carbon concentration still has its original value  $F$ .

The ratio between the quantity  $M_\alpha$  of the alpha mixed crystals (ferrite) separated at the end of the time  $t$  to the quantity  $M_\gamma$  of the residual  $\gamma$  mixed crystals (austenite) is expressed by the inverse ratio of the two segments  $\overline{E_2F'}$  and  $\overline{F'F_2}$ .

$$\frac{M_\alpha}{M_\gamma} = \frac{\overline{F'F_2}}{\overline{E_2F'}}$$

Indicating with  $M$  the entire mass of the metal under consideration,

$$\frac{M_\alpha}{M - M_\alpha} = \frac{\overline{F'F_2}}{\overline{EF_2'}}$$

The quantity of ferrite of average composition  $E_2$  which will be separated from austenite after the time  $t$  at the temperature  $T_1$ , will therefore be the smaller, the nearer the point  $F_2$  will be to the point  $F'$ . After all, this is self-evident, since  $F_2$  is the average composition of the residual austenite and  $F'$  is its initial composition.

Point  $F_2$  is moved toward the left and nearer to the point  $F'$ , as the amount of diffusion becomes less. At the same time it is also evident that the quantity of alpha mixed crystals separated at temperature  $T_1$  will be larger, the more completely carbon has diffused throughout the residual gamma mixed crystals. Naturally, the tendency will be for carbon to diffuse through the solid solutions from the high-carbon interior-contact with carbon-poor ferrite, outward into the surrounding austenitic material of original and medium concentration.

Especially we see that the proportion of ferrite in the metallic mass will increase, if the temperature  $T_1$  be kept constant for a time.

In this investigation we should also take into account the disturbances arising from the fact that similar diffusion processes take place in the carbon of the alpha mixed crystals contemporaneously with the diffusion of carbon in the austenite. But the study of these disturbances would immensely complicate the investigation of the process with which we are concerned. Nor, on the other hand, would it present any practical interest, because the amount of carbon diffusing in austenite is so far greater than in ferrite that the effect of the latter disturbance is practically negligible. This lack of proportion between the importance assumed by the phenomena occurring in the two solutions is due to the greater lineal velocity of diffusion of carbon in austenite, other conditions being equal, as well as to the fact that both the absolute value and the differences of the carbon concentrations in ferrite are far below those in austenite.

**51.** Let us now see how the allotropic transformation  $\gamma \rightleftharpoons \alpha$  progresses when the temperature continues to fall below  $T_1$ , reaching  $T_2$  after another time interval  $t$ .

supersaturated with iron in comparison to the theoretical equilibrium with the alpha mixed crystals. Supersaturation of that austenite also increases in amount with each decrease in temperature. From this it follows in accordance with the well known properties of supersaturated solutions, that step by step as the temperature of the system is falling, the tendency grows toward the formation of new crystallization centers in that austenite preserving the initial concentration  $L$ .

This much established, it is clear that during the interval of temperature between  $T_1$  and  $T_2$ —which is supposed to have been traversed by all parts of our system at the same speed during time  $t_1$ —the allotropic transformation  $\gamma \rightleftharpoons \alpha$  develops in three distinct ways:

(a) Around ferrite of average composition  $E_2$ , separated during the preceding interval of temperature, will form a new layer of mixed crystals of the same series, in which the carbon concentration at the external surface will reach the maximum value  $E_3$ . Contemporaneously, diffusion of carbon inward from the periphery of each of the ferrite crystallites formed in the second short lapse of time will have increased the carbon concentration at their center. As a result, the average carbon content of the new ferrite masses formed in this way around the original nucleus will be represented by a new point  $E_4$  situated at the right of  $E_2$  but at the left of  $E_3$ .

At the same time the austenitic regions surrounding each mass of ferrite described immediately above has also increased in carbon content, reaching the maximum value  $F_3$  at the surface of contact between the gamma and alpha phases.

Simultaneously the thickness of the shells of austenite partially enriched in carbon and surrounding the same ferrite crystal has also been increased as a result of the diffusion of carbon from the interface between phases back into the solid solution, a diffusion rendered more intense by the increase of the maximum concentration of carbon from  $F_1$  to  $F_3$ .

In regard to the ratio between the entire mass  $M$  of the part of our system under consideration and the mass  $M'\alpha$  of the ferrite separated therefrom at temperature  $T_2$  around germs appearing at the start of the transformation, we may repeat the considerations made in reference to the previous interval of the same process. Thus, indicating with  $F_4$  the point representing the average composition of the  $\gamma$  phase at temperature  $T_2$  (at the end of the second short time interval  $t_1$ ) we may establish the following relation:

$$\frac{M'\alpha}{M - M'\alpha} = \frac{\overline{F''F_4}}{E_4\overline{F''}}$$

In regard to the position of the point  $F_4$ , the same remarks already made in reference to the corresponding point  $F_2$  apply (page 67).

(b) In that part of the austenite supersaturated in iron, in which the carbon concentration has remained unchanged at the original value  $L$  on account of the incomplete development of the diffusion processes, new centers of crystallization will form during the second minute period of time  $t_1$ , around which in turn new masses of ferrite will gather. The formation of alpha mixed crystals of this new series will proceed in a manner similar to that of the previous alpha crystals formed around the nuclei appearing in the former interval of temperature  $T - T_1$ . But it is clear that, at the end of the second interval of time  $t_1$ , the crystals of the new series will have reached their normal growth—corresponding to the temperature  $T_2$  of the system—in a great deal shorter time than the older ones. As a consequence, the younger ferrite growing in that part of the austenite which has not yet been enriched in carbon by its diffusion from the regions surrounding the older ferrite of the first series will have greater differences in carbon concentration edge to center. Consequently the mass of the ferrite crystals separated at a given temperature from a given metallic mass, by virtue of non-simultaneous appearance of crystallization germs, is yet smaller than the mass of ferrite which would have separated had complete diffusion of carbon taken place so as to insure total homogeneity in each phase at all times. This follows as a corollary from the relation established in the preceding paragraph.

(c) New centers of crystallization may be born at any time in that heterogeneous austenite immediately surrounding the alpha mixed crystals. True, the carbon concentration in these regions has an equilibrium percentage only at the surface of contact with the ferrite, but in all the remaining mass, the austenite is supersaturated with iron. Ferrite crystals appearing in this region may be denoted by group three. The process of formation of group three is characterized by occurrences intermediate between the corresponding values for the transformation  $\gamma \rightleftharpoons \alpha$  in the other two regions examined under (a) and (b) immediately above. In other words the proportions of the mass of two phases in reciprocal equilibrium at each instant will be less than the relative mass of the oldest ferrite, but greater than that of the youngest. The same is true for the variation in carbon concentration within the heterogeneous phases existing in equilibrium, metastable at best.

**52.** Having reached this point, before going further in the examination of the transformation process, it is necessary to explain clearly the significance of the elements of our diagram.

When studying the course of crystallization in molten steel in Part I, it was possible to limit consideration to the phenomena occurring *around a single crystallization-center*. Since in that case the concentration of all the residual liquid could be considered homogeneous at any instant due to the high velocity of diffusion, it is clear that the process was the same for



of the austenite. In quantity, this undercooled portion generally far outweighs the transformed metal during a large share of the first part of the transformation.

When discussing crystallization of molten steel, undercooling (or surfusion) did not appear as a factor, once solidification had started, because experience shows that in industrial conditions it takes place to a negligible measure, merely on account of the homogeneity easily maintained in the liquid phase.

Instead, undercooling is precisely the cause of the partial crystallization of  $\alpha$  iron in the three groups of crystalline individuals specified in the preceding paragraph.

These points established, it is clear that it is impossible further to limit the consideration of the changes accompanying allotropic transformation of the iron to the phenomena occurring around but a single crystallization-center of  $\alpha$  iron, since this is not the exact picture of the average process occurring in the entire metallic mass taken into examination. Consequently one is always obliged to state explicitly in each case whether the compositions represented by the various points of a certain equilibrium diagram refer to a single crystalline individual of  $\alpha$  iron formed around a given crystallization center (as we have done, for instance, in discussing case *a* in the previous section), or else whether those values refer to the average concentration in the entire mass of a given phase existing at a given instant, even when this mass includes various centers of crystallization of different age.

53. Observe, then, the course of the reversible transformation  $\gamma \rightleftharpoons \alpha$  diagrammed in Fig. 12 as it takes place within a finite interval of temperature  $T$  to  $T_m$  long enough to permit checking the theoretical conclusions with experiments on the phenomena which take place in the system.

As a basis will be taken what we have previously described as happening in the elementary intervals of temperature  $T$  to  $T_1$ ;  $T_1$  to  $T_2$  . . .  $T_{m-1}$  to  $T_m$  comprising the total interval  $T$  to  $T_m$ . In order to fix our ideas, we will consider the difference between the manner of separation of ferrite from solid solution austenite, and the manner of separation of austenite from the mass of the molten steel, such as has been studied in Part 1.

Perhaps it is superfluous to point out again that a widely different diffusion velocity of carbon is the first cause responsible for the divergent phenomena in the two cases. Carbon, indeed, as well as the other elements, but especially carbon diffuses from points of high concentration to points of low concentration throughout all parts of both

the phase stable at higher temperatures and the phases stable at temperature below the transformation point. Yet the velocity is a great deal higher in gamma solutions crystallizing out of molten steel than in the case of the transformation  $\gamma \rightleftharpoons \alpha$ . However, in reality, the immediate difference between the phenomena characteristic of the two cases is due essentially to the difference in the velocity of diffusion of carbon in the liquid melt from the velocity throughout the austenite just above  $A_r$ , that is to say, to the difference in diffusion velocity in the phase stable at higher temperatures. In fact, it has already been pointed out several times in Part I that upon solidification this velocity should be considered so high as to produce at each instant a substantially complete homogeneity in the liquid. Now such a situation never takes place even approximately in austenite, the phase stable at higher temperatures during transformation  $\gamma \rightleftharpoons \alpha$ .

It follows that in the latter case it is necessary to introduce entirely new elements into the equilibrium diagram to represent the various carbon concentrations characteristic of the various regions of the high temperature phase; elements not needed in the study of crystallization. As to the low temperature phase, however, nothing need be added to the conception of a band of variant concentration developed in Sec. 11 *e.s.* Only quantitative variations in its shape and location will be produced by the greater tardiness of diffusion in the alpha mixed crystals, stable at the lower temperatures.

54. The first new element which appears in the diagram as a consequence of incomplete diffusion of carbon is the locus of the points, denoted by the letters  $F_2, F_4 \dots F_m$  in Fig. 12. From the very definition of these points, it appears clear that the curve which they locate is nothing else but the "curve of average carbon concentration" in the untransformed  $\gamma$  phase remaining at each given instant during the  $\gamma \rightleftharpoons \alpha$  allotropic change.

It is opportune to examine briefly what the course of such curve should be during the entire interval of transformation.

As a matter of fact, the curve in question has a precise significance only upon condition that the lines and points of the diagram have reference to the carbon concentrations in the entire mass of steel taken under consideration. This may be emphasized immediately, although it is a necessary consequence of the discussion in the preceding pages.

It is evident when we observe that the locus of curve  $F_2 F_4 \dots F_m$  is determined not only by the speed of cooling but also by the frequency of the new centers of crystallization in the residual austenite

is moved more toward the left, away from the curve of the maximum concentration of carbon in the same phase, the greater the number of the new crystallization-germs continually appearing during the progress of the transformation.

55. But, in addition to this curve,  $F \dots F_m$ , one must also take into consideration a new line in the equilibrium diagram, namely, the curve of minimum carbon concentrations in the  $\gamma$  phase stable at the higher temperatures. It is clear that a similar curve could not be a distinct line in the equilibrium diagram in the region where liquid steel crystallizes. Since the carbon concentration can be considered as practically uniform in all parts of the liquid—the solution stable at the higher temperatures—both this last curve as well as the curve of average concentration will coincide with the liquidus of the system (curve  $F F_1 F_3 \dots F_n$  of Fig. 12).

However, for the case which is now being studied, it is clear that in the first steps of the  $\gamma \rightarrow \alpha$  transformation, the curve of the minimum carbon concentration in the  $\gamma$  phase coincides with the vertical representing the original composition of the system. Thus, during the first two time intervals examined in the previous paragraphs, when only an extremely small amount of material had been transformed, the vertical  $L F F' F''$  represents the carbon concentration in the untransformed austenite in those regions described in (b) of Sec. 51 in which the effect of diffusion has not yet been felt. Evidently this is also the minimum concentration of carbon existing in any of the true gamma phase, since that portion near the alpha solid solution has been enriched by migration of carbon from that austenite which upon transforming produces the ferrite low in carbon. We shall see shortly that this coincidence between ordinate  $L/L$  and the curve of minimum carbon in austenite occurs only during the first part of the total transformation range.

Finally, a third line in the equilibrium diagram for allotropic transformation differs markedly in significance from the corresponding one in the solidification diagram. This element is the curve  $A F F_1 F_3 \dots F_n$ , Fig. 12, denoting the beginning of transformation. It is the true liquidus when denoting the solidification of a binary iron:carbon alloy. Its points represent the uniform concentration of carbon in the liquid phase at the various temperatures. On the other hand, when denoting the transformation of gamma iron into ferrite, the points of such a curve represent at each temperature the carbon concentration at the interface where the  $\gamma$  phase is in contact with the alpha mixed crystals. Contrary to what has just been stated to obtain for the curve of average concentrations, the curve  $A F F_1 F_3 \dots F_n$  clearly has a precise significance even when it is applied to the transformation  $\gamma \rightarrow \alpha$  taking place around but one center of secondary crystallization.

completely homogeneous, the same curve  $F \dots F_n$  will represent the maximum concentration of carbon or saturation point as it varies with temperature and carbon content.

56. We have thus defined the principal elements of the diagram representing the course of transformation in the gamma phase. There are also three other similar curves for the  $\alpha$  phase stable at the lower temperatures, not differing *qualitatively* from their homologues which we have marked out during crystallization of molten steel. The former differ from the latter only for the *quantitative* displacement sideways caused by a given variation of the velocity of cooling. This is true in sense and in measure corresponding to the fact that, other conditions being equal, the velocity with which carbon diffuses is lower in ferrite than it would be in austenite.

Summing up the data which has been deduced, a list is appended of the principal lines in the diagram whose significance has been determined in the present and the previous chapters of this treatise, briefly mentioning their characteristics:

A. Curves representing the concentrations of carbon in the phase stable at the higher temperatures:

(a) Curve of minimum carbon concentrations in the  $\gamma$  phase. During a first period of the  $\gamma \rightarrow \alpha$  transformation it coincides with the vertical ordinate denoting the initial composition of the homogeneous system (line  $LF'F''$  in Fig. 12.) It has a precise significance only when considered as applying to the entire mass of untransformed  $\gamma$  phase which remains in the system at each instant.

(b) Curve of the average carbon concentrations in the  $\gamma$  phase (line  $FF_2F_4 \dots F_m$  in Fig. 12.) This curve also has a well-defined significance only when referred to the entire mass of the  $\gamma$  phase remaining in the system at each instant of the transformation.

(c) Curve of the maximum carbon concentrations in the  $\gamma$  phase (line  $FF_1F_3 \dots F_n$  in Fig. 12.) It corresponds to the curve representing the beginning of the transformation  $\gamma \rightarrow \alpha$ , when the concentration in all parts of the initial phase is practically uniform. It has a well defined significance only when we consider the process of the allotropic transformation  $\gamma \rightarrow \alpha$  going forward around only one center of crystallization.

B. Curves representing the concentration of carbon in the phase stable at the lower temperatures:

(a) Curve of the minimum carbon concentrations in the  $\alpha$  phase (line  $EE_o$ , Fig. 12.) Due to the small diffusion velocity of carbon in ferrite, this curve very nearly approaches a vertical, and in practice it can be considered as such. From what we have seen in the previous paragraphs, it is clear that it cannot have a precise significance unless it be considered in relation to the  $\alpha \rightarrow \alpha$  transformation proceeding around

(b) Curve of the average carbon concentrations in the  $\alpha$  phase (curve  $E E_2 E_4 \dots E_m$ , Fig. 12). It has the same characteristics of the corresponding curve marked out in the diagram representing solidification of molten steel. Note the important fact that a given reheating will cause a much smaller displacement of curve  $E E_2 E_4 \dots E_m$  than in the corresponding line in the austenitic field (see Sec. 20 and following). Again this is a consequence of a low velocity of diffusion at temperatures near the critical. This curve has a well defined significance both when considering the entire  $\alpha$  phase, as well as when considering only that portion formed around a single nucleus.

(c) Curve of the maximum carbon concentrations in the  $\alpha$  phase (line  $E E' E'_2 E_1 E'_4 E_3$ , Fig. 12). This line also has the same characters as the corresponding curve studied in the case of solidification. It has also a definite significance both when the entirety of the  $\alpha$  phase is considered, as well as when one takes into account only that particle of ferrite which has separated around a given center of crystallization.

## COMPLETION OF SECONDARY CRYSTALLIZATION OF UNIFORM AUSTENITE

57. Taking as a basis the principal elements of the binary diagram thus defined, it is not difficult to form an approximate idea of the course of the various phenomena when the system passes through the entire transformation range. It would not be so easy to determine in a quantitatively accurate way each step of the transformation for the various types of crystalline individuals forming each of the two phases, and so construct a precise equilibrium diagram. But inasmuch as such determinations would be superfluous for the practical aims in view, I deem it unnecessary even to mention the efforts which have been made from this angle, and of those lines of research which may be developed further in several directions.

With the lowering of the temperature, it is clear that the  $\gamma \rightarrow \alpha$  transformation will continue in the same manner and with the same mechanism which has been analyzed for the first two minute intervals of temperature  $T$  to  $T_1$  and  $T_1$  to  $T_2$ . This is true for those regions of the  $\gamma$  solid solution surrounding each crystalline element of the  $\alpha$  series<sup>1</sup> in which the concentration of carbon due to the incomplete development of the process of diffusion has reached values higher than the original one and yet the diffusing carbon has not invaded the entire mass of the residual  $\gamma$  phase.

Instead, the development of the phenomena undergoes some modifications from the moment when diffusion extends to the supercarburated regions around the  $\alpha$  crystals, and the same regions are multiplied in correspondence with new crystalline individuals continuously appearing in the residual austenite, causing the entire mass to be somewhat enriched in carbon. Evidently at that time the last trace of uniform  $\gamma$  solid solution disappears, such as was described in Sec. 51, (b), and in which the carbon concentration constantly remained the same as the initial composition of the  $\gamma$  phase.

From this point on during further cooling, it is evident that the curve of minimum carbon concentrations in the  $\gamma$  crystals [see Sec. 56, A, (a)] cannot further coincide with the ordinate corresponding to the original concentration of carbon in the homogeneous  $\gamma$  phase. Instead, the curve gets farther and farther away to the right as the temperature falls. In Fig. 13 is marked the course of this curve in  $ABC$ , supposing that the

<sup>1</sup> See Sec. 51, (c).

last residuals of  $\gamma$  solid solution of the original concentration ( $AB$ ) disappear at temperature  $T_m$ .

Although it is obvious after what has been said in Sec. 56 apropos of the meaning of each of the curves of the diagram, it should be remembered that the diagram in Fig. 13 is to be understood as referring to the concentrations of the *entire* mass of each one of the phases which are present in the system at each instant.

Starting from the moment in which the last portion of the  $\gamma$  solid solution of concentration  $AB$  disappears, it is just as clear that the course of the curve of the average concentrations of the  $\gamma$  phase [see Sec. 56, A, (b)] shall undergo a modification, due to the interruption of phe-

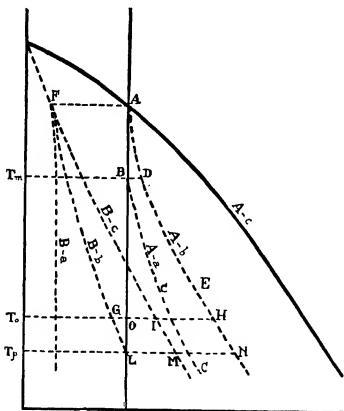


Fig. 13.—Complete decomposition of massive austenite.

nomena defined in Sec. 51, (b); phenomena of crystallization which, with equal intensity of the accompanying thermal phenomenon, would cause the smallest possible subtraction of carbon from the  $\gamma$  phase. In fact the  $\alpha$  phase which separated as a nuclear point at the birth of each one of these new crystalline individuals always contains the lowest carbon concentration (Fig. 12,  $E$ ) even when the process in question starts at a temperature much lower than that of the beginning of transformation of the system studied.<sup>1</sup> Therefore, it is easy to understand that when this group of phenomena stop, the fact must bring in consequence an increase in the quantity of carbon subtracted from the austenite caused

is in comparison to the quantity of carbon, which an equivalent variation of the thermal state of the system removed into the same  $\gamma$  solution in the previous time interval during which that earlier group of phenomena continued to exert its effects.

Let us represent with  $ADE$  in Fig. 13 the thus defined course of the curve of average carbon concentrations in the  $\gamma$  phase.

58. As a result of the very definition of the two curves  $ADE$  (average composition of the entire  $\gamma$  phase) and  $FGL$  (average composition of the entire  $\alpha$  phase), it is evident that at each temperature  $T$ , below  $A$ , the relationship

$$\frac{M_{\alpha}}{M_{\gamma}} = \frac{\overline{OH}}{\overline{GO}}$$

exists between mass  $M_{\alpha}$  of the  $\alpha$  phase separated from the system, and the mass  $M_{\gamma}$  of the  $\gamma$  phase.

The proportion between ferrite and austenite evidently does not depend exclusively upon the velocity of diffusion of carbon in the phase stable at the lower temperatures, as it did in the case of crystallization of molten steel, but it depends also upon the velocity of diffusion in the phase stable at the higher temperatures. After all, this resulted from the considerations set forth in the previous paragraphs, and has been developed in Sec. 51, (a), as it applies to a particular portion of the  $\alpha$  and  $\gamma$  phase.

In other words: during the solidification of molten steel the instantaneous proportion of the two phases may be derived merely from a curve corresponding to  $FG$  of Fig. 13, yet during the  $\gamma \rightarrow \alpha$  transformation the same proportion is expressed by the relations existing between the course of same curve  $FG$  and that of another curve  $DEN$ .

Finally, from what has been said in a previous chapter regarding solidification of molten steel, it is evident that the  $\gamma \rightarrow \alpha$  transformation will be completed at temperature  $T_p$ , at which point curve  $FG$  cuts the vertical  $AOL$ .

59. Having established these fundamental data, let us resume observing the actual course of the phenomena which more especially interest us at present in view of the practical purposes in view: namely, those phenomena which accompany the closing stages of the reversible transformation in a pure binary hypoeutectoid steel.

In Secs. 46 and 47 I have already briefly examined such a transformation taking place in materials whose average analysis falls without the limits of the  $\alpha$  solid solutions. However, the considerations developed in the last few pages now allow the study of the  $\gamma \rightleftharpoons \alpha$  change in the way it really happens in practice. Previous to this time we were obliged to work under the restriction that diffusion in the phase stable at higher temperatures (austenite) must be rapid and complete, since we lacked the



Figure 14 contains merely the essential elements denoting the course of an actual transformation of initially homogeneous  $\gamma$  mixed crystals.

There is again found curve  $AB$ , the average composition of the  $\gamma$  phase [Sec. 56,  $A$ , (b),] and curve  $CD$ , the average composition of the  $\alpha$  phase [Sec. 56,  $B$ , (b)]. Suppose that the first series of  $\alpha$  mixed crystals (that is to say, that part of the  $\alpha$  crystal containing the lower percentage of carbon) is saturated with carbon at temperature  $T_1$ . Let  $F$  be the carbon concentration corresponding to this saturation. At temperature  $T_1$ , therefore, the region of insolubility or of heterogeneous equilibrium (Italian, *lacuna di miscibilità*; German, *Mischungslücke*) in the  $\alpha$  phase will extend from the point  $F$  to the right, reaching a point  $H'$  which in

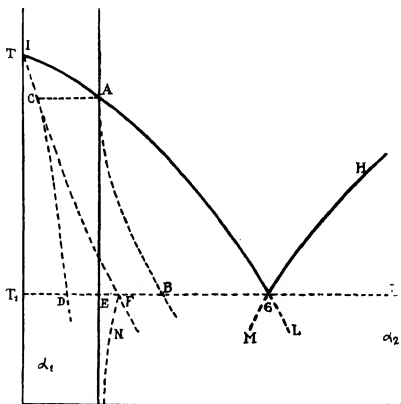


FIG. 14.—Outline diagram at transformation range.

Fig. 14 falls outside the sketch. It is known that the point  $H'$  corresponds to pure cementite  $\text{Fe}_3\text{C}$  (6.67 per cent. C) in the special case of steel, now being studied. Any iron-carbon alloy with a carbon concentration falling between  $F$  and  $H_1$  on arriving at  $T_1$  as  $\gamma$  solid solution will separate on further cooling into conjugate solutions  $\alpha_1$  and  $\alpha_2$ —two distinct series of crystals.

Let us now see how the transformation  $\gamma \rightarrow \alpha$  proceeds when the temperature of an alloy  $AE$  is lowered to  $T_1$ .

Once this temperature is reached, it is clear that the maximum concentration of carbon cannot further increase in the saturated

To be more precise, once a temperature slightly below  $T_1$  is reached, the maximum concentration of carbon in the  $\gamma$  crystals will have grown a small quantity above the value  $G$ . The point representing the composition of this phase will have moved along the curve  $GL$ , a prolongation of the liquidus  $IAG$ . It is clear that at such a moment the system is undercooled and in a state of metastable equilibrium, being supersaturated in respect to  $\alpha_2$  phase.

However, after a small lapse of time (whose duration changes with the general conditions of the system, and with the velocity of cooling, but, which at any rate, is always very small) the undercooled state comes to an end; a certain quantity of the  $\alpha_2$  phase (in our case cementite of composition  $H'$ ) separates abruptly with the evolution of latent heat so that the temperature of the mass increases suddenly, returning to  $T_1$ . From this point further abstraction of heat from the system causes a continued separation of the  $\alpha_2$  phase along the curve  $GM$ , and thus the condition of metastable equilibrium returns, in this case the mother solid being supersaturated with  $\alpha_1$ . When this undercooling reaches its limit, new quantities of the  $\alpha_1$  phase (ferrite of slightly variable composition in accordance with curve  $FN$  as concentration varies with temperature) separate abruptly with evolution of latent heat until the temperature of the alloy has again moved up to  $T_1$ .

As heat is withdrawn from the system, the process of alternate separation of small quantities of the  $\alpha_1$  and  $\alpha_2$  phases continues. This is the mechanism of formation of the "eutectoid" structure, in steels known as "pearlite," an aggregate of minute portions of the  $\alpha_1$  solid solution (ferrite), alternating with small elements of the  $\alpha_2$  phase (cementite).

60. I would not have stopped to describe in this brief and not rigidly accurate manner the mechanism of the formation of the eutectoid  $G$  were it not that I needed to clearly fix in the reader's mind the position of the points and lines in Fig. 14 representing various phases of the process.

Now let us see how the concentrations of carbon in the eutectoid will vary during its formation.

In the first place, our special point of view permits us to concern ourselves only with variations in composition of the  $\alpha_1$  constituent, inasmuch as experience shows that the second phase,  $\alpha_2$ , separates as a practically pure compound and, therefore, with a practically constant composition  $H'$ , corresponding approximately to the formula  $\text{Fe}_3\text{C}$ .

It also follows from what has been said immediately above that the eutectoid will form only in those regions of the  $\gamma$  phase in which the carbon concentration has reached the value  $G$ . In fact, it will start along the vertical line  $GL$  (formed in the interval  $TT_1$ ) and

$\alpha_2$  (cementite) will be reached at a later time, the more rapid the decrease in carbon concentration when one passes from the initial  $\alpha_1:\alpha_2$  interface into the layers of the solid solution contiguous. Consequently, the proportion of ferrite (more strictly the  $\alpha$  solid solution of variable concentration denoted by the curve  $NF$ ) in the eutectoid will be greater, the steeper the carbon gradient existing at the instant of first formation of the eutectoid. Even when the formation of the eutectoid is more or less advanced the same apparent excess of ferrite will be produced if the carbon concentration rapidly decreases in the material immediately surrounding the already separated pearlite toward the layers of the same  $\gamma$  solution farther away from the external surface of those nuclei.



Fig. 15.—Pearlite showing large segregation in ferrite.

But the indicated variation of carbon concentration in the material surrounding a given crystallization-center of  $\alpha$  phase is the less abrupt, the earlier such nucleus has formed in the history of the transformation  $\gamma \rightleftharpoons \alpha$ . This is because the time during which the processes of diffusion have had an opportunity to take place has been greater, and has rendered that variation less abrupt, and at the same time has started at higher temperatures at which the intensity of dispersion is greater.

From this it follows that—while in practice the carbon analysis of the eutectoid pearlite always remains approximately constant and close to that represented by point  $G$ —the ratios between the quantities of its two structural constituents may appear under the microscope to undergo local variations, depending upon the different manner in which the dif-

fusion of carbon proceeds. This phenomenon in turn is bound to the occurrence of undercooling accompanying the  $\gamma \rightleftharpoons \alpha$  transformation, and to the frequency of the  $\alpha$  nuclei born in the steel.

These observations explain the irregularities and seeming abnormalities which can be almost constantly detected in the pearlite constituents.

Figures 15 and 16 show two examples. When considered in the light of the remarks set forward in this section, they are clear enough by themselves so that it is superfluous to add any explanation.

61. Having thus examined the character of the eutectoid formed by the special conditions of diffusion supposed to hold for our system, let us briefly note how the transformation of the entire system proceeds. It will be noted that at temperature  $T_1$  the system has become invariant



FIG. 16.—Pearlite showing large variation in texture.

due to the appearance of a new phase,  $\alpha_2$ , and when we continue to abstract heat from it, the temperature will remain constant until the  $\gamma$  phase disappears.

With this end in view, again observe the distinction between the three different regions of the  $\gamma$  phase, described in Sec. 51 as being those in which the  $\gamma \rightleftharpoons \alpha$  transformation occurs in different ways, and discover

have already outlined the characteristics of this new transformation in the preceding section.

(b) In region (b)—characterized by a carbon concentration substantially unaltered from  $AE$ , the original analysis of the system—the transformation proceeds in an absolutely identical manner to that already studied in Sec. 51 for the same region. All the observations made there apply to this transformation which takes place, by the way, in an under-cooled system. It will cause a gradual sidewise displacement of the point  $B$ , representing the average composition of the residual  $\gamma$  phase, along the horizontal  $T_1BG$  from  $B$  toward  $G$ .

(c) In third region (c)—immediately surrounding region (a) and possessing a variant carbon concentration which analyses greater than that of region (b)—the transformation  $\gamma \rightleftharpoons \alpha$  proceeds in the same way described in Sec. 51 when studying the corresponding region. Here the transition also causes a movement of the point  $B$  along the eutectoid horizontal toward  $G$ , as explained in Sec. 51.

From definition of the region (b), and from what has been said in Sec. 57, it is clear that at a certain instant of time that region must disappear. Thereafter the transformation proceeds according to the two schemes peculiar to the remaining regions (a) and (c). It frequently happens also that region (b) has finally disappeared at a temperature above the eutectoid temperature  $T_1$ . Naturally, the entire eutectoid transformation then takes place according to the manner peculiar to (a) and (c). In this case the ratio between the mass of the  $\alpha$  mixed crystals actually separated in one of the regions (b) or (c) and the mass of the  $\alpha$  crystals which would have separated had there been total homogeneity in each phase, varies with each change of frequency of crystallization-germs.

It is clear that one cannot further use considerations based upon the concentrations of the various elements in each phase when attempting to derive the ratio existing between the quantity of pearlite and that of ferrite which will constitute steel of composition  $AE$  (Fig. 14) when its temperature has lowered beneath the eutectoid temperature  $T_1$ . Deductions based upon composition are valid only when the system is in complete equilibrium so that the composition of each phase is constant in all its parts. Nor can we start with an analysis based upon the *average* composition of the individual phases, because that reasoning could apply only when all the particles of the second phase may always be derived from the first one under the same conditions. Such a contingency was discussed and accepted in Part I because the liquid metal existing in equilibrium with solid austenite during the mushy stage

previous sections, it clearly results that such a state never happens when austenite breaks up at  $A_r$  ranges.

In order to deduce some positive indications even of a generic character from those same considerations in addition to the previous negative conclusion, let us briefly recount the conditions under which the eutectoid transformation is completed.

It has already been seen that during the  $\gamma \rightleftharpoons \alpha$  transformation now being studied various reasons concur in producing displacements of point  $B$  toward the right. From the very definition of this point it is clear that it cannot be thrust in that direction beyond point  $G$ . On the other hand, it is known that transformation  $\gamma \rightarrow \alpha$  takes place at temperature  $T_1$  after the manner of a stable system within those regions in which the carbon concentration in the  $\gamma$  phase has the exact value  $G$ . In the other regions of the heterogeneous mass it occurs as the breakdown of a metastable state. But even in these undercooled regions, the transformation has always been accompanied since the arrival at  $T_1$  by the formation of eutectoid austenite (concentration  $G$ ). It is therefore evident that at the end of the process the last residue of  $\gamma$  phase will necessarily have a carbon concentration equal to  $G$ . Hence at that time point  $B$  will have reached position  $G$ .

It may be superfluous to add that at the end of the eutectoid transformation the same point  $G$  will then become the limiting position of the intersection between horizontal  $T_1G$  with curve  $BC$  of Fig. 13, representing the minimum carbon-concentrations in the  $\gamma$  phase.

Point  $D$  also moves toward the right during the eutectoid transformation. But, from the various observations previously made, especially in the end of Sec. 51 and Sec. 60 and 61, the amount of this perturbation and its consequent end-point depends upon the velocity of cooling, upon the mean distance between the alpha nuclei, and upon the importance of the undercooling which accompanies the formation of the new crystallization-germs in the different regions of the  $\gamma$  solid solution.

If we admit as a sufficient approximation that the composition of the eutectoid is represented by the point  $G$ , the same ratio between its mass and that of the  $\alpha_1$  mixed crystals existing when the eutectoid transformation is completed, is equal to the ratio between the distance  $\overline{DE}$  and the distance  $\overline{EG}$  (Fig. 14). It is therefore clear that all the causes which produce a right-hand movement of the location of  $D$  also produce a decrease of the proportion of the eutectoid in comparison with  $\alpha_1$  mixed crystals, and *vice versa*.

In addition to this, however, it is necessary to remember the remarks previously emphasized that the transformations which take place at temperatures below  $T_1$  in slowly cooled steels proceed with a velocity so small that in practice they can be totally neglected.

## CHAPTER IX

### SECONDARY CRYSTALLIZATION OF NON-HOMOGENEOUS AUSTENITE IN THE BINARY IRON : CARBON SYSTEM

63. In the study of the transformations which the binary iron:carbon alloy undergoes during cooling subsequent to its complete solidification, it has formerly been supposed<sup>1</sup> that the austenite has been maintained at a high enough temperature for a sufficient lapse of time so that the diffusion of carbon would produce perfect uniformity. Let us now see what modifications result when we lift the restrictions imposed by the above hypothesis, which in practice—as I have already said many times—never occurs, not even approximately.

In order to do this, one is obliged to consider the composition of an austenite crystal not as being represented by a "point," but instead by a line, a portion of an abscissa drawn to each successive temperature. Moreover, it has been demonstrated in Part I that the integration of such lines delineates a "band of variant concentration."

Call  $AB$  (Fig. 17) the range of carbon concentrations in the  $\gamma$  solid solution of an alloy whose average composition is  $H H'$ , as it exists at some temperature  $T$  below the solidus ( $CD$ ) but above the initial point of the  $\gamma \rightarrow \alpha$  transformation.

The band of variant concentration which must be taken into account when the temperature of the system has fallen below  $T$ , is bordered by two descending curves  $a$  and  $b$  which gradually approach one another. The distance between the two curves at a given lower temperature  $T_1$ , is less, the slower the cooling from  $T$  to  $T_1$ . Now let us see how the  $\gamma \rightarrow \alpha$  transformation takes place in a system thus constituted. In order to visualize it in the clearest manner, we shall begin by restricting our examination to the phenomena which happen within the sphere of *only one primary crystalline element*.

The laws upon which this restriction must be based have already been noted.

We know that in such a case, the points of the curve  $a$  (Fig. 17) correspond to the concentrations of carbon at the center of the crystalline element under examination, while curve  $b$  denotes the concentrations of carbon at the periphery of the same element.

When the system has reached temperature  $T_2$ , at which the curve  $a$  cuts  $EF$ , the liquidus of the  $\gamma \rightarrow \alpha$  transformation, crystals will start to

<sup>1</sup> See Sec. 46 and 47.

by the nucleus of primary crystallization.

Here is an essential difference between the course of the phenomena in the more general and real case now under examination and in the occurrences which would appear in perfectly homogeneous solution, already studied in the previous sections. While in that first ideal case any necessary relations between the position of the initial centers of crystallization of the first-appearing solid solution and the centers of primary crystalli-

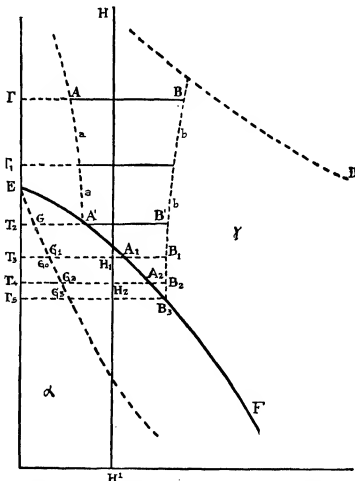


FIG. 17.—Decomposition of non-uniform austenite.

zation were not observed, in an actual case such a relation appears immediately and in the simplest form: the two points must coincide.

A more complete study of the phenomena which I now wish to present in their simpler schematic form, would show that in practice the difference between the two cases is never clearly defined. This is because even in the ideal case, the presence of impurities, which are never possible to eliminate totally in steel produced in ordinary practice, establishes a relationship between the network of the primary crystalline structure and that of the secondary, even when the  $\gamma$  solution has been brought to the highest degrees of homogeneity which can be attained in



industry. In the case of ordinary steel the presence of the same impurities may so strongly modify the process of secondary crystallization as to prevent the identity of position of the initial centers of this re-crystallization and of the primary solidification, although it does not eliminate (as it is easily understood) but rather accentuates the other relations obtaining between the two crystalline systems. For simplicity, the effects produced by these complementary phenomena will not be taken into account at present, though it must be realized that their practical importance may approach and sometimes greatly exceed that of the fundamental phenomena now under study, especially when very impure steels are under observation.

64. When after a very short lapse of time  $t$ , the temperature of the system has dropped further, passing from the value  $T_2$  to the value  $T_3$ ,<sup>1</sup> a small mass of  $\alpha$  solid solution will form around the center of transformation defined in the preceding section. The carbon in this mass will have an average analysis lying between  $G$  and  $G_1$  and its specific concentration will vary from the center toward the periphery of the transformed mass from a minimum value corresponding to a point  $G_0$  situated slightly at the right side of  $G$  to a maximum value equal to  $G_1$ . The reasons for all this have been seen in Part I.

At the end of the same interval of time  $t$ , the carbon content in the region surrounding this ferrite mass, especially at the surface of contact between the residual  $\gamma$  solution and the external surface of the small mass of  $\alpha$  solution, has increased to the value  $A_1$ .  $A_1$  corresponds precisely to the conditions of equilibrium between the  $\gamma$  mixed crystals and the  $\alpha$  crystals at temperature  $T_3$ .

The same principles already indicated at the beginning of Sec. 50 allow two portions of the residual austenite to be distinguished at the end of the time  $t$ :

(a) The first fraction forms something like a shell around the mass of  $\alpha$  solid solution which precipitated in the temperature interval  $T_2$  to  $T_3$ . As just remarked, its carbon concentration has the maximum value  $A_1$  at the internal surface of contact. Outwards it diminishes rapidly in the successive thin layers. The external limit of region (a) is marked by the layer in which the concentration of carbon has remained unaltered from that which it possessed before the ferrite nucleus had separated. It is clear that the thickness of this "shell" or, in other words, the depth to which the  $\gamma$  solid solution has been enriched in carbon by the separation of the first accumulation of  $\alpha$  crystals, other conditions being equal will be greater, the greater has been the velocity of the diffusion of carbon, or—for a given value of the velocity of diffusion—the longer has been the time  $t$  required for cooling the system from the temperature  $T_2$  to  $T_3$ .

<sup>1</sup> The proportions between the different parts of the diagram are out of scale, as

(b) The second portion of the residual  $\gamma$  solution is that constituted by all the remaining mass of austenite, surrounding the region defined in (a). In this region, where the effects of diffusion of the carbon accumulated in zone (a) have had no time yet to be felt, the concentration varies from a minimum value (located between  $A'$  and  $A_1$ ) in the layer corresponding to the external limit of region (a) and the maximum value  $B_1$  at the external layer of region (b), a layer which naturally coincides with the external surface of the primary crystalline element which is under observation.

In regard to the quantitative ratio between the  $\alpha$  mixed crystals formed at temperature  $T_3$ , and the residual austenite then remaining unaltered, I could now repeat, except with a few perfectly obvious modifications, what I have already said in Sec. 51, again reaching the conclusion that the quantity of ferrite will be the larger, the more complete the diffusion of carbon in the residual austenite.

65. Contrary to what was the usual occurrence in the preceding discussions,<sup>1</sup> this last conclusion is true for the non-uniform austenite under consideration only when the time interval  $t$ , as well as the temperature interval  $T_2$  to  $T_3$ , is very small. In fact, one must also take into account the diffusion of carbon in the austenitic region (b) of Sec. 64, a region which in previous discussions had a uniform carbon concentration. Here the diffusion of carbon progresses in a contrary direction to that in which it proceeds in region (a); i.e., inward from the external layers of the primary crystalline element. It is easy to understand that this inward migration counteracts the effects of the analogous process taking place in region (a), the amount of the outward diffusion from ferrite crystallites and the effects originating from it.

Among these last effects, one must recall the one which I mentioned at the end of Sec. 50, namely, the influence of cooling velocity upon the proportion of  $\alpha$  mixed crystals. Particularly it must be clear that the constant tendency of heterogeneous austenite to become uniform in carbon concentration counteracts the process by which  $\alpha$  mixed crystals can be proportionately increased by slowing the cooling or prolonging the heating.<sup>2</sup> In the more general case of heterogeneous austenite now under observation, the inward diffusion in zone (b) may reach such intensity with large differences in carbon concentration as to annul and sometimes even invert the effects which reheating the steel would be expected to produce upon the proportion of  $\alpha$  mixed crystals separated from the  $\gamma$  solid solution at a given temperature.

Here is a first example of the marked structural differences which must occur in steel, according to whether the history of the piece has caused a minor heterogeneity in the primary  $\gamma$  solid solution (approaching

<sup>1</sup> Especially see Sec. 50 and 51.

<sup>2</sup> See the last lines of Sec. 50.

the system cools. The temperature has reached the value  $T_4$  after a second short lapse of time  $t_1$ .

Again it is convenient to consider separately three distinct regions in the residual austenite:

(A) is the peripheral region of the primary crystalline element under observation throughout which the concentration of carbon is to the right of the liquidus  $EB_2$  at temperature  $T_4$  (Fig. 17). It is bounded by the external surface of the crystalline element, where carbon has a concentration  $B_2$ , and the surface at which the concentration has the value  $A_2$  exactly corresponding to a point on the liquidus. At  $A_2$  austenite is evidently saturated in respect to the  $\alpha$  solid solution at the temperature  $T_4$ .

From observation of the diagram, transformation  $\gamma \rightarrow \alpha$  does not take place in any point of region (A) at temperatures above  $T_4$ . But in this region, diffusion of carbon from the peripheral layers toward its innermost layers (Sec. 65) is continually in progress during all the time the metal is above temperature  $T_6$ .

(B) is the region immediately around the  $\alpha$  mixed crystals formed in the preceding minute time interval required to cool from  $T_2$  to  $T_3$ . Its size is determined by the distance to which the influence of the ferrite crystallite built up about the original nucleus may be felt. Separation of new quantities of  $\alpha$  solid solution during the temperature interval  $T_3$  to  $T_4$  takes place in it without undercooling. The newly born ferrite also forms a new layer around the crystallite formed in the previous interval  $T_2$  to  $T_3$ . In this new layer however, the concentration of carbon at its external surface will reach the maximum value  $G_2$ . But at the same time, the process of diffusion will have raised the concentration at the central region of the same ferrite individual by transporting carbon inward from the periphery. As a result, the average carbon concentration in the enlarged  $\alpha$  crystal will be represented by a new point  $G_3$ , situated at the right of  $G_0$  but at the left of  $G_2$ .

At the same time the carbon has also increased in the  $\gamma$  solid solution surrounding this new layer of  $\alpha$  solid solution, reaching the maximum value  $A_2$  at the interface. And contemporaneously the dimension of Zone (B) [which in Sec. 64 was defined as shell (a) of austenite] has increased due to the diffusion of carbon in the  $\gamma$  solid solution occurring within the second time  $t_1$  and rendered more intense by the concomitant increase in saturation with further cooling, represented by the change from point  $A_1$  to  $A_2$  (Fig. 17).

This increase is also counteracted by the same phenomena studied

which also forms a part of the shell enveloping the latter. It is entirely limited on the outside by the surface at which the carbon concentration has the value  $A_2$ ; a surface which, as has already been seen, limits the inside of region (A).

From the definition already given of region (B), it is clear that the internal limits of (C) cannot be indicated with equal precision, since it is determined only by the distance beyond which the action of the first particles of ferrite as a "crystallization-germ" does not make itself felt. From all that was said in the previous paragraphs, it is clear that in region (C) the  $\gamma$  solid solution is supersaturated in respect to  $\alpha$  mixed crystals. Therefore, in general, new crystallization centers will form in this region around which new crystalline elements of the  $\alpha$  series will develop.

Owing to the schematic character of the diagram upon which our study is based—it would not be wise to affirm that this last phenomenon actually takes place in the second time interval  $t_1$ , rather than being initiated in a successive interval of time. But inasmuch as the course of the process continues identically in the manner just described until new centers of ferrite crystallization appear in region (C), it is clear that an argument will lose nothing in generality if, for simple reasons of brevity, it is supposed that the appearance of new germs actually occurs in the time  $t_1$  which we now are studying.

It is obvious that the phenomena accompanying the growth of such new crystalline elements of ferrite in region (C), take place under conditions entirely similar to those accompanying the growth of the  $\alpha$  crystallites of the first series, formed at the central region of the primary austenite crystal. The difference between the elements of the two series is purely quantitative and consists in the fact that the  $\alpha$  mixed crystals last to appear are born in a solution of higher carbon content, so that their nuclei are also higher in carbon.

A second quantitative difference is found in this last fact, whose effects accrue to those studied at the end of Sec. 65. Consideration will show that the consequences of transformation in region (C) counteract those of the phenomenon explained at the end of the previous section, as far as concerns the proportion of  $\alpha$  mixed crystals separated at a determined instant in an alloy of given initial composition. As formerly limited, all the new crystalline elements which formed during a phase of the transformation corresponding to the one now under observation occurred in the region around the high carbon "shell" enveloping  $\alpha$  crystals of the first series—region (b) of Sec. 51—and were necessarily obliged to show a minimum carbon-concentration equal to that of the crystals of the previous series.

It is clear that the ratio between the carbon concentrations of the two phases,  $\alpha$  and  $\gamma$ , and, as a consequence, the ratio between the quantities of the same two phases which are reciprocally in unstable equilibrium at a given instant of the transformation, are subject to further variations due to the different velocities at which the various transformation and diffusion processes take place. This is evidently true both for homogeneous and for heterogeneous austenite. An analytical observation of these variations would not only take too long, but it should also presume at least an approximate knowledge of the relations between the diffusion velocity of carbon and the temperature of the system; also between the frequency of the ferrite nuclei, the amount of undercooling in the austenite when such germs are born, and the temperature of the system.

The same observations apply in this regard which I outlined in Sec. 51, (b).

67. Now let us observe—as we did in Sec. 53—how the reversible transformation progresses when the system cools to a temperature below  $T_5$  (Fig. 17).  $T_5$  is the intersection of liquidus  $EF$  with curve  $b$ , the right border of the band of heterogeneity in the  $\gamma$  solid solution. It is also necessary in this analysis to bear in mind the observations contained in Sec. 52.

It is clear that, in all the phases of the process which occur at temperatures above  $T_5$ , the trend of the various phenomena does not differ qualitatively from that which we have briefly analyzed for the first phases of the transformation during the minute time intervals  $t$  and  $t_1$ . Note that Fig. 17 is badly out of scale, in that  $T_4$  to  $T_5$  represents a considerable temperature interval, while  $T_2$  to  $T_4$  is extremely small.

Having reached the temperature  $T_5$ , the iron-carbon alloy is in a similar state to that of the same alloy at the same temperature in case the original austenite had a uniform concentration of carbon.<sup>1</sup> The differences have already been pointed out with a brief analysis of their causes and characters and consist only in the distribution of the secondary ferrite and in the carbon content. As a mere consequence of this last fact a different proportion exists between the mass of the two constituents themselves.

But this certainly does not modify the further course of the transformation studied in the previous pages. It is easy to understand this point if the considerations which I have made from Sec. 53 up to Sec. 62 are observed from this special point of view. To justify this observation now would only result in repeating what was said there.

In order to avoid misunderstanding as to the generality of the phenomena described in Sec. 53 to 62, it should be pointed out that it is perfectly possible to extend those principles to a mass of steel containing any number whatsoever of primary or stalline elements. In fact, a

simple revision of the wording will make the text cover a sizable quantity of steel, maintaining the value of the remarks intact even when applied to an assemblage of any number of primary crystalline elements. The original analysis was restricted to the case of only one crystal for reasons of simplicity and clearness. However in this more general case it is necessary to consider the total quantities of each portion of the two constituents  $\alpha$  and  $\gamma$ , which at each period in the process of transformation appear with the specific characteristics already studied. Also it is necessary to consider the average carbon concentrations in the total mass of said distinct portions.

One could profitably repeat the observations made in Sec. 54 *e.s.* regarding the location of the curves representing the average and minimum concentrations of carbon in the  $\gamma$  phase, with only this difference: that in the present case this curve may only accidentally coincide with the ordinate representing the average concentration of carbon of the alloy, while in the previous case this coincidence always occurs in a determined interval of the transformation.

The specific characteristic of the curve discussed in Sec. 55 representing the points where  $\gamma \rightarrow \alpha$  transformation begins for each determined carbon concentration applies to the present discussion for temperatures below  $T_5$ . This curve will not represent the maximum concentration of carbon in the entire  $\gamma$  phase except for temperatures below  $T_5$ .

It is also easy to see that the observation made in Sec. 56 concerning the curves denoting the carbon concentrations in the  $\alpha$  phase can be applied to the general case which we are now studying; also those of Sec. 57 and 58 in reference to the relations between the velocity of diffusion of carbon in austenite, and the proportion of the  $\alpha$  and  $\gamma$  phases co-existent in the system at each instant.

Sections 59 and 62 discussed the limiting case where austenite, originally uniform in carbon concentration, decomposed at the eutectoid horizontal into a conglomerate of  $\alpha_1$  and  $\alpha_2$  phases (ferrite and cementite). In general, this former discussion applies to the more usual industrial condition where heterogeneous austenite is transformed. In applying the data of Sec. 60 regarding the anomalies in the formation of pearlite due to slow diffusion of carbon in austenite, it is necessary to take into account what was said in Sec. 65 concerning the influence of heterogeneous austenite upon the carbon content of the ferrite nuclei at the time of their birth, and also upon the ratio between the co-existent volumes of the two regions ( $B$ ) and ( $A$ ).

68. The schematic considerations already made up to this point would allow one to deduce some interesting practical consequences of the variations imposed upon the structural characters of a hypo-eutectoid steel by the conditions of its solidification, of its subsequent cooling and of the heat treatment to which it is subjected after its complete solidifica-

tion. However, inasmuch as these structural variations are intimately connected with changes in the mechanical properties of steel, I deem it opportune to postpone its study to the successive parts, in which I shall briefly examine some examples of the effects which heat treatments may produce in the mechanical properties of the metals comprising our investigation.

Before leaving the study of the binary iron:carbon alloy to pass to more complex steels such as really appear in practice, I will only add a remark, negative in character, which avoids the necessity of taking under observation a series of phenomena, whose analysis would involve great complications and difficulties in our study.

If one observes the whole of the iron:carbon equilibrium diagram comprising all the steels, properly speaking (namely, up to a content of 1.8 per cent. carbon), it may be seen how it is possible for some primary cementite to form even in hypo-eutectoid alloys, especially if containing but little less than 0.9 per cent. carbon. Referring to Fig. 11, it is clear that if the band of variant concentration in the  $\gamma$  mixed crystals of average analysis even below that of the point  $H$  is so large as to extend toward the right beyond the point  $H$  at the eutectoid temperature  $IGH$  ( $690^{\circ}\text{C.}$ ), then a portion of the transformation  $\gamma \rightleftharpoons \alpha$  will take place in the richest austenite during cooling in correspondence to the points of the curve  $HN$ , and will cause the separation of primary cementite.

The process just mentioned explains one of the causes producing the paradoxical appearance repeatedly observed by many experimentors, of contemporary pro-eutectoid ferrite and cementite in hypo-eutectoid carbon steels. In some cases the same fact originates from a different cause, connected with the abnormal development of the processes of undercooling, seen in Sec. 59 to occur in the formation of the eutectic.

This phenomenon does not happen in the great majority of that special class of machine steels which forms the object of our study. In those few cases in which it does occur it never attains noticeable importance either upon the microstructure of the alloy or upon the mechanical properties. It is, therefore, quite justifiable to disregard it in a summary study.

69. Although the observation presents a direct interest only in the application of thermal analysis to the study of our steels, and is not apt to be utilized for technical purposes, it is also opportune to point out that for an iron:carbon alloy of a given average composition, the "liquidus" of the reaction  $\gamma \rightarrow \alpha$  (thermally determined as the locus of the points at which, in cooling, the reaction now indicated starts) will appear to be moved toward the upper and the right side, the wider the band of variant concen-

## CHAPTER X

### SECONDARY CRYSTALLIZATION IN MORE COMPLEX SYSTEMS

70. Let us see now, how the course of the phenomena studied in the previous paragraphs can be graphically represented for the only cases which really appear in practice, namely, those in which the steel considered is no longer a binary alloy, but contains one or more of the other metals constantly entering into the composition of the ordinary machine steels.

In other words, let us lift the third restriction which we adopted in Sec. 46 and 47 when beginning the study of the  $\gamma \rightleftharpoons \alpha$  transformation. The first and second of these restrictions were eliminated in Sec. 59 and 63 respectively.

In order to study this general case completely and intently, one ought to examine again the entire considerations already established in Part I regarding primary crystallization, in order to show how the various elements of the "real" diagrams in three dimensions have to be modified at the transformation range. This would show precisely how the so-called "fictitious" binary and ternary diagrams, representing the more complex systems, are derived from accurate sketches representing binary and ternary equilibria.

However, such a method would be too lengthy, and would necessitate a large number of superfluous repetitions. Therefore, it is preferable to discuss these "fictitious" diagrams which in the simplest way represent the course of the transformations of our system completely enough to suffice for all practical purposes. A few remarks will be added when necessary to establish the characteristics of the new lines and areas in them which must be considered. Especially this will be necessary to describe the position which these new markings assume in comparison with the one they occupy in true diagrams in three dimensions. Anyone who wishes to follow more precisely and closely in its details the phenomena following the  $\gamma \rightleftharpoons \alpha$  transition in a more complex steel, has only to review the pages of Part I, taking into proper account the subject matter of the previous sections of Part II. He will easily realize the fact that all the other possible considerations beyond those before mentioned regarding the fictitious transformation diagrams in two and three dimensions, and their derivation from the true diagrams, are identical to those already developed in part I in the case of primary solidification, or may be easily



the concentration of the various elements entering into the steel.

Furthermore, as has already been done in Part I when studying solidification, and for the same reasons advanced in Sec. 33 and 34, carbon and an iron complex will be taken as principal constituents, representing the carbon concentrations as abscissæ of a fictitious binary diagram.

All those conditions which apply to the elements which with carbon enter into the composition of our steels, apply to the  $\gamma \rightleftharpoons \alpha$  transition much the same as they affect the process of primary crystallization. In Sec. 32 to 34 are to be found some observations necessary and sufficient to justify the deduction of a simplified "fictitious" diagram from complete diagrams too complicated to be directly applied to practical uses. Such remarks also apply to the  $\gamma \rightarrow \alpha$  transformation.

Although the experimental data available is in reality very scarce and not always definite, it warrants the statement that the course of the  $\gamma \rightleftharpoons \alpha$  transformation in some ways appears a great deal more suitable for representation by methods described in Part I and there applied to the study of solidification. This is due to the fact that the transformation range for the binary iron alloys with the constituents of our steels beside carbon are always narrower than the intervals of the primary crystallization. Therefore the "*deformations*" of the "true" binary or ternary diagrams (by means of which we pass to the corresponding "fictitious" diagrams of the more complex alloys) are generally smaller.

This happens in an especially remarkable measure for all the constituents other than carbon which enter into alloy with iron forming the so-called "carbon machine steels." In such steels, these constituents are ordinarily limited to silicon and manganese, beside minute quantities of the usual impurities (sulphur, phosphorus, copper, arsenic, etc.) Normally such impurities exist in extremely low percentages in machine steel, never much exceeding 0.2 to 0.3 per cent. Silicon and manganese also rarely go above 0.30 per cent. and 1 per cent. respectively.

For these steels therefore, there cannot be any doubt that the best method of graphic representation is the fictitious binary diagram, with iron and carbon assumed as basic elements.

If instead, we pass to the so-called "alloy steels," we find almost always in them an element beside iron or carbon preponderant both for its quantity as well as for its specific effects. And in this case—as we have already seen for the primary crystallization—it would generally seem best to adopt a fictitious ternary diagram, having as basic elements iron, carbon and this element X.

However, the fact has been mentioned before that the transformation ranges are rather small for binary alloys of iron with one of the metals

which frequently enter into the composition of special machine steels such as nickel, chromium, tungsten, molybdenum, or vanadium. Consequently it is convenient to adopt the simpler fictitious binary diagrams for representing the greater number of those steels. Another reason is that the “*deformations*” caused even by a rather high proportion of one or more constituents are not such as to confuse this method of representation.

For these reasons, I will refer in what follows only to the case of the fictitious binary diagram.

Besides, it is very easy to extend the conclusions which we shall reach in regard to the binary diagram to the case of the fictitious ternary diagram, basing the work upon what we have seen in Part I.

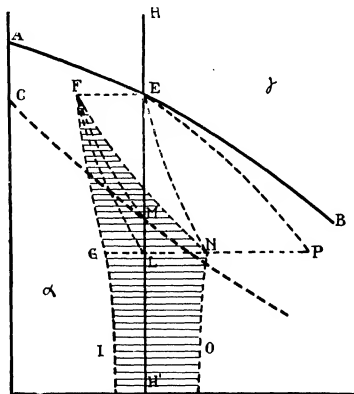


FIG. 18.—Decomposition of uniform austenite in a complex steel.

72. If curve  $AB$  in a fictitious binary diagram Fig. 18 represents the beginning of the  $\gamma \rightarrow \alpha$  transformation and  $CM$  represents its ending, it will be easy here to delineate the elements indicating the progress of the phenomena examined in Chapters VI to IX.

First assume for the sake of simplicity that uniformity in concentrations has been reached and will be constantly maintained in the  $\gamma$  solid solution. One may immediately assemble the experimental data concerning the transformation  $\gamma \rightleftharpoons \alpha$  in the diagram in the following principal lines:

*First.*—The curve  $FGI$ , denoting the minimum carbon concentrations in the  $\alpha$  mixed crystals.

*Second.*—Curve *FL* representing the average carbon concentrations in the same mixed crystals. *L*, its point of intersection with the vertical *HH'*, corresponds to the end of the transformation.

*Third.*—*FN* is the curve of maximum carbon in the ferrite. Its intersection with the horizontal through *L* is the point *N*, determining the concentration of carbon in the last traces of  $\alpha$  mixed crystals to be formed. Therefore, it marks the lower extremity of the curve *EN* followed by the "fictitious representative point" of our system as defined in Sec. 17.

*Fourth.*—Curve *EP* represents the concentrations of carbon in the  $\gamma$  mixed crystals at various temperatures during the transformation range.

*Fifth.*—The two curves *GI* and *NO* limit the band of variant carbon concentration in the  $\alpha$  solid solution.

From a practical point of view it would evidently be of no special interest to determine the present position of the curve *EP*, in comparison to the position which the same curve would occupy if diffusion in ferrite were speedy and complete.

However, the same observation is not true for the three curves *FG*, *FL*, and *FN* in regard to their position with relation to curve *FM*, where as has been demonstrated, all these three curves would meet if complete equilibrium of concentrations in ferrite could be reached in all stages of the transition. In fact the temperature of the end of transformation and the width of the band of heterogeneity in ferrite depend upon the position of the curves.

Reflection will show that the reciprocal position of curves *FN*, *FG* and *FL*, in respect to curve *FM*, cannot be anything but the one indicated in Fig. 18. Naturally, their divergence from the location of *FM* will depend upon the nature of the alloy considered, the conditions under which solidification has taken place, and upon the subsequent cooling velocity of the solid system. As already remarked, this may be thoroughly established by again examining the principles laid down in Part I regarding the method of passing from the "true" diagrams of simple systems to those "fictitious" ones corresponding to the more complex.

**73.** If now we analyze the  $\gamma \rightarrow \alpha$  transformation, taking into account the fact that the velocity of diffusion of carbon and of the other elements in the  $\gamma$  phase is never sufficient to cause complete homogeneity, we see immediately that Fig. 18 contains all the elements necessary to apply to this more complex case all the reasoning of the previous paragraphs concerning the  $\gamma \rightleftharpoons \alpha$  transformation in the binary iron-carbon alloy. On repeating those considerations we will again find the three new curves defined in Sec. 55 and 56 representing the concentration of carbon in austenite, as it varies with temperature and location. We may also

plied to the study of the more complex real system. This statement applies both in regard to the simple uninterrupted transformation  $\gamma \rightarrow \alpha$ , and to the instances where the effects of the insoluble range in the  $\alpha$  mixed crystals are felt, resulting in the formation of peralite.

However, it is necessary to remember that in all cases, but especially in the phenomena connected with eutectoid formation, the variations of concentrations of *all* the elements due to the incomplete diffusion give rise to new effects which reinforce those directly due to imperfect carbon diffusion. These last effects have been studied in the preceding pages only in the binary iron: carbon alloy and we have heretofore only considered the possibility of graphic representation of incomplete diffusion in our fictitious binary diagram. It is evident that the ternary diagrams studied in Part I would also contain the possibility of serving as a basis for graphic representation for these more complex phenomena. Such a study would be very long and complicated, and at present the quantitative experimental data are totally lacking which alone could give it some practical interest. It is clear that such missing data are just what would serve to construct the equilibrium diagrams of the binary systems formed by iron and each constituent of machine steels. It would be necessary to determine the value of the disturbances caused in such diagrams by the incomplete development of diffusion of each of those elements.

Perhaps it is superfluous to point out that the effects of slow diffusion are felt in the fictitious binary diagram (Fig. 18) by sidewise displacements of the six fundamental curves defined in Sec. 55 and 56. However, the experimental data now available are insufficient to furnish indications, not only about the amount but even about the direction of these perturbations.

Experience actually shows that in many ternary and quaternary special steels, the displacements above indicated reduce the width of the band of variant concentrations in comparison with the dimensions which the latter reaches in plain carbon steels of equal carbon percentage, other conditions being equal.

**74.** Finally, in order to eliminate the last restriction which has been imposed upon our schemes of graphic representation, it remains to investigate the case of a complex steel used for the construction of machine parts in which the course of the primary crystallization and of the succeeding cooling of the resultant  $\gamma$  solid solution has not been so slow as to produce complete uniformity of the various elements in the same solution.

In order to sum up the conclusions which we have reached, it is convenient to observe the entire fictitious binary diagram, including the interval of solidification of liquid steel as well as the  $\gamma \rightleftharpoons \alpha$  transformation range. Such a diagram, reproduced in its essential parts in Fig. 19, does not contain any new elements in addition to those studied in the pre-

Both in Fig. 17 and 18 the same letters have been used to indicate corresponding points.

Suffice it to point out that the curve  $OP'$  is the curve representing the maximum concentrations of carbon existing in the mixed crystals which still remain in the system at the various temperatures comprised in the same interval.

It is superfluous to note that the same markings contained in the diagram of Fig. 19 could easily be modified so as to cover the more general

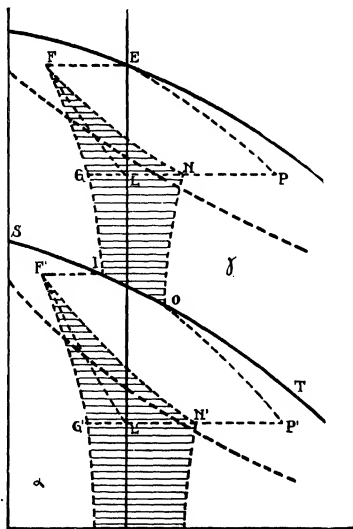


FIG. 19.—Constitution of complex steel during slow cooling.

case regarding the formation of eutectoid pearlite; it would only require an extension of the considerations developed in Sec. 59 to 62 and 67 to 69.

75. A most general study of the effects which certain definite heat treatments may produce upon the transformation might now be based upon the ideas developed in the previous pages. These heat treatments would always consist in determined variations of the velocity of cooling of the metallic alloy, sometimes alternated with reheatings, or holdings at a given temperature for some time (soaking).

However, such a general investigation would not present a real interest unless it was illustrated with an accompanying detailed and complete study of the modifications in physical and mechanical properties outlined above which accompany variations in the course of the transformations. But, under such conditions, it is clear that such an investigation could not find an adequate place in this volume both because of its magnitude and its character.

Therefore, I plan to present later only a relatively few practical examples illustrating how the previous deliberations can be employed both as an aid in the interpretation of the phenomena occurring in practice, by deriving from them precise indications of how to modify their course rightly, and as a guide in searching for a heat treatment more suitable to impart the properties thought to be the best for the use of a given steel.

In the following few pages I will only add some observations concerning two classes of phenomena which must be taken into account in the best practical applications of the principles derived in the previous sections.

Both classes of these phenomena may be thought of as being "retards" in the start and in the development of the transitions. Therefore, in a cooling steel the transformations start at a lower temperature than would be expected from scaling the diagram of "normal" equilibrium. They would also proceed but incompletely. Published equilibrium diagrams are obviously based on experimental data determined under such conditions as to avoid the intervention of these very phenomena.

As I have already said, they may be subdivided in two principal groups:

(a) Retardations due to the fact that the given system can be cooled considerably below its transformation temperature, yet at the same time remain entirely unaltered. In other words, it is then in a *metastable* condition possessing the form normally stable only at a higher temperature.

Such phenomena occur in the  $\gamma \rightarrow \alpha$  transformation as well as in the primary solidification. In the later case they assume the characters of true "undercooling" phenomena, and may cause effects of the greatest practical importance, such as, for instance, great accidental segregations or liquations.

When phenomena of this group occur in the field of allotropic transformation, they may spread so much that the metastable system reaches temperatures so low that the transition takes place only at an extremely small speed. Sometimes the metastable state cannot revert to the stable; in this case we pass directly to retardations of the second group (b), below, whose effects can be connected directly to those of the first.

Retardations due to undercooling are characterized, as is well known

an action is similar to that caused by "primers" on an explosive, and is instituted by the appearance of "germs" of the phase stable at lower temperatures. If the reaction is "detonated" at a sufficiently high temperature to allow the transformation being completed at an observable velocity, once started, it continues with a development of heat until the system is converted to a condition of stable equilibrium.

(b) Retardations of a given transformation upon cooling due to the fact that it normally starts and takes place in a range of temperature so low that the velocity with which it proceeds is not sufficient to render the change noticeable in short intervals of time.

Phenomena of this group appear with greater intensity in special steels with high percentages of alloying metals, as can be naturally expected if we remember the equilibrium diagrams of the complex steels. Other conditions being equal, their variations due to changing the nature and concentration of the various metals constituting the particular special steel, can be established with certainty *a priori*, when the equilibrium diagram of said steel is known in its more complete form studied in the previous pages. We have already seen clearly how the phenomenon can be represented graphically, when we studied the criteria by which we pass from the graphic representation of the simple systems to that of the more complex systems by the introduction of the "fictitious diagrams."

This fact now constitutes indeed one of the properties which differentiate retardations of group (a) from those of (b). The former are not represented by any of the markings of the diagrams which we have studied.

A second differentiation lies in the fact that recalescence naturally cannot appear in the latter.

It is superfluous to point out that the retards belonging to this second group pertain not only to the continuous transformation  $\gamma \rightarrow \alpha$ , but also happen when the transformation generates pearlite. This appears evident from what we have seen in the previous sections.

Here it is necessary to observe that, in practice, the two orders of retardations almost never appear separately. In general, those of the first group (a) appear frequently and to a greater measure in steels of low normal transformation points. Thus, for these steels, it almost always happens that when retardation due to undercooling ceases, the metal is already at a temperature at which the transformation velocity is extremely low.

76. Other conditions being equal it is well known that the amount of retardation is greater for speedy cooling. Again, the facility in reaching temperatures at which an uncompleted transformation proceeds at unobservable rates permitting no noticeable segregation of new phases, increases with higher velocity of quenching.

It is also known that advantage of this fact is taken in the practice of what I call "quality" or "final" heat treatments, designed to impart determined mechanical properties to a given steel by means of a correct selection of the velocity of final cooling.

Hardening without drastic quenching is effected in the well known processes of air hardening, hot oil hardening, etc. Such processes are based upon these very criteria, reinforced by thermal hysteresis occurring in greater measure in the special steels of low transformation intervals. With the aid of those processes it is possible to obtain the same effects by means of simple reheatings of steels of a suitably selected composition, which upon ordinary steels can be obtained only by means of drastic quenchings, an expedient which would often be disastrous.

This is not the place to discuss the phenomena and processes just mentioned because their study should more appropriately be a part of a treatise on "quality" or "final" heat treatments than of a study of "homogeneity" or "preliminary" heat treatments, which are based essentially upon the utilization of the phenomena studied in the previous pages.

At any rate, I wanted to mention them here, because they are sometimes taken advantage of in the "preliminary" heat treatments. Their existence must be known and noted, even if use is not made of their special characteristics. I propose to return later to this point in reference to some concrete examples.





## PART III

# DIFFUSION IN AUSTENITE AS APPLIED TO THE PRELIMINARY HEAT TREATMENT OF STEELS

## CHAPTER XI

### FUNDAMENTAL CRITERIA FOR PRACTICAL APPLICATION OF PRELIMINARY HEAT TREATMENTS

77. The principles developed in the proceeding pages permit tracing the fundamental lines of a schematic chart in a broad manner. So-called preliminary heat treatments for homogeneity (normalizing) are based upon the principal phenomena represented by such a chart.

Our experimental knowledge which can actually be accurately placed in such a sketch is very scarce and often uncertain. At present it can therefore be stated that merely the first elements of a systematic doctrine exist which if known would permit the deduction of practical rules for the preliminary heat treatment of a steel of a given chemical composition by the mere application of criteria directly inferred from that schematic chart. An enormous mass of detailed experimental material must be gathered and classified before the necessary theoretical studies can fully synthesize a diagram in such accurate particulars as to render it susceptible of easy, direct, and general applications in steel practice. From what has been seen in the previous sections, it appears clear that this enormous analytical experimental work shall comprehend all the problems, without exception, which I have been able only to mention during our summary and rapid examination. It would be superfluous to repeat a list of such problems ranging all the way from the accurate determination of all the possible binary equilibrium diagrams which may be formed by the elements entering into the composition of our steels, to the determination of the laws of diffusion of each one of these elements in the alloys in question.

From what has been briefly said in the preceding chapters, it may easily be seen that the successful issue of such researches will require the use of a very great variety of experimental methods: ranging from thermal analysis to microscopic examination; from the purely chemical methods to the methods based on the study of electrical and magnetic properties of the alloys, and so on. Nor is it at all unlikely (as recent researches seem to prove) that new investigative methods, founded on principles which have never yet been used in such field of studies, may prove of

78. Since the deficiencies and the uncertainties are so great which render many parts of the proposed diagrams incomplected and undetermined, one might imagine that the considerations made up to this point would be entirely useless, constituting nothing more than a pure theoretical exercise. This is not my opinion, and it may be useful to mention briefly my reasons.

As has been said there were two purposes aimed at in compiling these notes—nor could they possibly have been different, for the reasons which are now evident at this point of our study.

The first was to show how it is possible to rationally coordinate the meager experimental data now possessed. In my opinion, this is the best method to indicate, at least in its main lines, what may be the more desirable path for further experimental researches. Especially the light of past knowledge is necessary in order to draw results from fresh researches useful for the final purpose and to make them susceptible of easier and surer interpretation. So far as this first aim is concerned, I could have added nothing else to what has been said in the first two parts of this study. The actual state of our knowledge would not permit the theoretical investigation to be extended much further nor developed in greater detail, without totally abandoning the possibility of checking the results experimentally.

I, therefore, consider that the present state of our knowledge has been reviewed as completely as the limits of an elementary treatise permit with the presentation of the considerations developed in the previous chapters.

But my second aim depends a great deal more upon my conviction of the utility of the study previously developed in outline.

This second aim was to place the real mechanism of the complex phenomena which are utilized as a basis for the preliminary heat treatment in clearer light by defining its nature with that precision which only a graphic representation permits.

The observations made in the first two parts might also be sufficient upon this second point had the reader thoroughly known the technical nature of normalizing practice and had precise ideas regarding the nature and importance of the variations in structure and physical properties which may thereby be produced. Only in such an event might it be possible for those observations to reach fully their ultimate purpose. Then the connection between numerous elementary experimental facts could be discerned each one of which was previously known as an isolated occurrence. A more precise idea of the mechanism of the more complex phenomena would also be obtained by an appreciation of the laws according to which the effects of various elementary facts accrue one upon the other.

However, it does not seem logical that the main practical aim of a

treatise, elementary, or better preliminary, in character, should be exactly understood in its real value only by those who are already conversant with the phenomena to be coordinated. It seems especially illogical since this practical knowledge is apparently not widely scattered. Instead, beginners in this special branch of metallurgy should become familiar with the real practical value to be had from such a schematic study. Such knowledge will much facilitate the exact understanding of processes whose direct technical study can furnish nothing else but confused and indetermined ideas, even though this was the only method of attack possible until a short time ago.

From such a standpoint, it is evident that the discussion in the previous pages is not complete, nor could it be sufficient to give a definite notion of its real practical value to the beginner.

The best way to fill this gap would be to use that material in the development of a systematic treatise, as complete as present knowledge allows, dealing with the phenomena occurring in the principal metallurgical processes ordinarily applied in making and working machine steels from tapping to the final cooling, at least for as many of those phenomena as enter in the schematic chart already drawn. However, such a research would necessarily be very lengthy and would lead far beyond the boundaries which we have definitely set for our study.

For all these reasons, it is more desirable to follow a middle course. Consequently I intend to add to the formerly developed analysis a certain number of practical observations, which taken as isolated examples are suitable to illustrate, even to the beginner, the real nature and the practical value of the phenomena which we have so far studied only in outline.

This may also encourage some reader to deepen and to coordinate his theoretical and practical knowledge in this very interesting field of metallurgy.

79. However, before investigating such examples individually, some considerations of wide scope may be pointed out, a knowledge of which is necessary in order to understand the real practical value of those examples. All these observations would easily find their natural place in the course of a systematic and complete study of our general problem. But, inasmuch as it is impossible to proceed with such a study for reasons already given, I am obliged to indicate them separately.

To many readers they will appear tiresome, but they may be useful to others.

80. The so-called "preliminary heat treatment" of a low or medium steel used for mechanical purposes demanding toughness<sup>1</sup> as the most important physical property has for its essential if not exclusive purpose to minimize the heterogeneity of composition between the various part of

the primary and secondary crystalline constituents. Such treatments are therefore often called "homogeneity" heat treatments or normalizing. When the maximum result of the preliminary heat treatment has been reached, the steel is also in the best condition to acquire the physical properties such as hardness, tensile strength, elastic limit, elongation, etc., required for its correct use, at the same time maintaining the maximum compatible tenacity.<sup>1</sup> Such additional attributes are acquired by means of "final" or "quality" heat treatments—largely quenchings followed by low temperature reheatings.

In the first two parts of this volume, it has been said that the causes of heterogeneity in chemical composition belong essentially to two distinct series.

The first includes liquation phenomena occurring during separation of austenite from the molten mass of steel during primary solidification. To the second, however, belong all the analogous phenomena which accompany the transformations of  $\gamma$  iron into  $\beta$  iron and  $\beta$  iron into  $\alpha$  iron when the temperature crosses the critical ranges.

In the previous general study of these phenomena it has been seen that the results of allotropic transformation depend closely upon the effects produced by the phenomena of the first series. Sections 63 to 75 demonstrated that the heterogeneity existing below the critical temperature varied directly with the heterogeneity remaining in just-solidified austenite, other things being equal.

Therefore, it is evident that the study of industrial applications of normalizing heat treatments must take into account all the phenomena depending upon conditions of solidification and of transformation, not only separately but also as the earlier affects the latter.

This will appear clearer after the forthcoming examples have been presented.

**81.** From what has been said in the previous section and from the discussion contained in the first two parts of this volume, it appears that the principal groups of data determining the details of the preliminary heat treatment and the results thereby obtained, are the following:

*First.*—Course of primary crystallization.

*Second.*—Method of cooling the mass of  $\gamma$  mixed crystals after total solidification.

*Third.*—Course of the  $\gamma \rightarrow \alpha$  transformation.

Each preliminary heat treatment can consist in nothing else but a modification of the progress of one or more of the phenomena pertaining to these three groups.

The field of practical application considered in its extension a

nomena accompanying solidification. From a theoretical point of view there can be no doubt of this, as we have seen in previous chapters the great influence which the course of primary crystallization has upon the heterogeneity which the steel maintains in all its subsequent history. Yet in ordinary steel making practice, the study and the control of all the technical processes aimed at regulating casting practice are very seldom part of the duties of the person who has in charge the studying and the prosecution of all the other steps in the heat treatment.

The current dimensions and type of ingot molds, teeming temperature, length of time the ingot remains in the mold before stripping, cooling of the ingot after stripping, etc., is with rare exceptions in exclusive charge of the steel maker. Its control is far away from the man who is responsible for the further heat treatment of the metal.

Here is not the best place to discuss the reasons for such a division of technical duties. Doubtless they depend essentially upon the peculiar necessities of individual works organization. At most it can be emphasized that this division of routine duties requires a perfect accord between the criteria applied in the two series of processes, ordinarily in charge of different men. Also it follows that the man who has charge of the ingot pouring, for instance, should have a good knowledge of the ideas which are being applied by the man in charge of the subsequent mechanical and thermal treatment. For instance, every time it is necessary to cast large masses of special steels, it is absolutely necessary to take the greatest care to prevent excessive wholesale segregation of its components. This is done daily, by the way, in the manufacture of armor-plates of great thickness, as well as of fittings and portions of large caliber guns.

At any rate, the fact remains that at the present day the things which happen to a steel during solidification are not considered as being part of the "heat treatment" proper. Therefore, in selecting the examples further on, I shall pass by those more directly concerned with the effects of a determined mode of primary crystallization, and the variations of the properties of the metal which it is possible to obtain by modifying this routine.

Besides, we have already seen that the nature of the phenomena of solidification does not differ essentially from that of the second subdivision, resulting from the method of cooling the austenite.

**82.** We have already seen that this second series of phenomena are of a very simple nature, consisting essentially in the formation of the band of variant concentration and in the variations of its size.

We will presume that the primary crystallization has been regulated in such a way as to minimize the dimensions of this band of heterogeneity existing at temperatures immediately below that of the end of solidification. This is the best result which may now be expected in practice.

inasmuch as up to the present time it has been impossible to find a process analogous to quenching. Later we are going to use quenchings to avoid or reduce the formation of heterogeneities of concentration arising from the  $\gamma \rightarrow \alpha$  transformation. An analogous proceeding might also find industrial application as a means to regulate the process of primary crystallization, suppressing the effects of liquation or segregation which usually accompany solidification. However, up to the present time all trials made in that line and known to the author have been a failure.

Having once obtained the minimum heterogeneity in just-solidified austenite, it follows from what has been seen in the previous chapters that the immediate problem of heat treatment as far as it is affected by the second phase of the process is concerned (*i.e.*, the life history in the austenitic region) is reduced to the determination of the more suitable conditions for a simple reheating. In fact in this phase of the process it is only a question of how best to facilitate and accelerate the diffusion of the various elements contained in solid solution in  $\gamma$  iron. These latter processes take place exactly in such a direction as to cause a decrease of the dimensions of the band of variant concentration.

The most favorable conditions for producing uniform masses of steel consist in maintaining the metal at the highest possible temperature as long as possible. However, in practice, the application of this simplified idea is subject to restrictions of the highest practical importance, some examples of which it may be timely to mention here.

A first group of restrictions in the use of prolonged annealings at very high temperatures includes the tendency under those conditions to a modification of the crystalline structure, dangerous to the physical properties of the metal.

It is not exactly clear of what this modification really consists. Generally it is thought to be due, at least in a large part, to a gradual rearrangement in the orientation of the crystalline elements in such a way as to facilitate the formation and the extension of cleavage planes. Another very important factor causing damage upon excessive annealing is the "growth" of the primary crystalline grains due probably to surface-tension.

It is not necessary to dwell upon this first group of facts in these pages because they represent purely crystallographic traits and they have nothing to do with the physico-chemical phenomena which interest us now. They also have a prevalent negative importance from the standpoint of the heat treater inasmuch as they constitute essentially a mere restric-

Up to the present time, it is impossible to give general rules regarding the time and the temperature beyond which a heating would cause a certain degree of damage. Therefore, this data can be obtained only by direct experiment upon each type of steel, teemed under determined conditions. Since it is so difficult to obtain precise data upon this point, it is usually preferred to maintain normalizing heats well within those limits of time and temperature indicated as dangerous by experience upon analogous steels.

83. A second important restriction to the use of high temperatures in homogeneity heats has already been mentioned in Sec. 25 and consists in the necessity of keeping below the temperature at which steel starts "burning."



FIG. 20.—Slightly burned nickel-steel.  $\times 60$ . Etched with alcoholic solution of picric acid.

It is well to point out immediately that in common parlance the expression "burned steel" is often used to indicate a steel which has suffered dangerous but obscure modifications mentioned at the end of the previous section due to an excessive heating. This use of the expression is not appropriate. Such a steel ought to be designated simply as "over-heated." It can be "restored" by means of a suitable heat treatment or hot-working and the previous qualities regenerated. A steel which has actually been "burned" is ruined and is good for nothing but remelting scrap.

The term "burned steel" is used correctly to designate a steel which after its complete solidification has been reheated until it has reached the temperature of the final point of solidification in those of its parts which solidified last *i.e.* where the concentration of the various constituents in



steel which has been "burned" cannot be restored by any mechanical or thermal treatment, short of remelting.

In Chap. III, Sec. 25 and 26, the mechanism of the phenomenon of "burning" of steel has been already investigated quite fully. Only two actual examples of the characteristics shown in such material will be added here.

Figure 20, at 60 diameters enlargement after etching with an alcoholic solution of picric acid, shows the microstructure of a 2 per cent. nickel steel only slightly burned during the heating subsequent to the first forging operation.

Nevertheless, the deterioration of the physical properties due to such incipient "burning" is very remarkable. Such a steel as the type reproduced in Fig. 20 will give the following variations in physical properties determined by tension test pieces cut longitudinally with the direction of the forging:

State of the metal	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent. (cylindrical test piece 13.8 × 50 mm.)	Reduction of area, per cent.	Appearance of the fracture
After first forging at correct temperature and reheating to 800°C.	89,300	54,000	24.0	37	Normal, fibrous
After incipient burning occurring during heating for the second forging.	45,500	Could not be determined	2.5	0	Coarse, shiny grains

Figure 20 shows clearly the tenuous films of non-metallic matter—oxides and silicates—formed in the peripheral regions of the primary crystalline grains by oxidation of those portions which have suffered incipient fusion.

When burning is more intense its effects upon the physical properties of the steel may become far more serious until it produces such a disintegration in the metallic mass as could be actually characterized by the word "crumbling." From what we have seen in Chap. III, such a state of affairs naturally occurs more easily in the higher carburized steels and the decay is due to the fusion of the intergranular matter, annihilating crystalline adhesion.

Figures 21 and 22 show one of these more serious cases occurring in a steel ingot containing 1.07 per cent. carbon and 2 per cent. chromium, charged by mistake into a heating furnace to ether with other ingots of

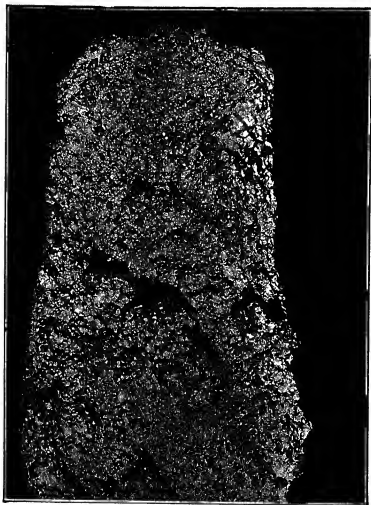


FIG. 21.—Fragment of crumbling overheated steel. Half size.



mild steel, and heated at the high forging temperature suitable only for the latter. The alloy ingot suffered true disintegration. Figure 21 at about half the natural size shows the appearance of one of the resulting fragments, and the "granular" structure assumed by the material may be clearly seen. The other Fig. 22 shows its microstructure at 100 diameters after etching with an alcoholic 5 per cent. solution of picric acid. Non-metallic inclusions may be clearly seen much more developed than in Fig. 20, where they were just forming.

It is possible to increase the minimum burning temperature of a given steel by means of preliminary heat-treatments, as has already been described with sufficient precision in Sec. 25.

## CHAPTER XII

### SELECTION OF PRELIMINARY HEAT TREATMENT ACCORDING TO THE CHARACTERS OF THE $\gamma \rightarrow \alpha$ TRANSFORMATION

84. As mentioned at the end of Part II, there are large classes of steels whose critical interval *on cooling* occurs at very low temperatures, for a number of reasons. The same steels almost always present the phenomena of thermal hysteresis to a great degree. A combination of these two properties sometimes reaches such a sum that not only the end of the allotropic transformation on cooling but also the point at which it is initiated falls below the minimum temperature at which such reaction proceeds with measurable velocity. In other words, sometimes the speed of cooling from above to below the transformation range as plotted on the equilibrium diagram may be so high that the velocity of transformation may be considered as practically zero. In such cases we have seen that, for cooling velocities higher than a certain value, the  $\gamma \rightarrow \alpha$  transformation does not advance in perceptible measure. It follows that such metal retains at ordinary temperatures that degree of homogeneity resulting from the conditions under which it solidified and from the subsequent homogeneity reheatings. Ordinary practice quite generally allows a simple cooling in air after a heating. Yet in such practice the principal factor or, better, substantially the only factor determining whether the allotropic transformation shall be completed or not, is the dimension of the different parts of the pieces treated.

To allow all natural transitions to occur in pieces of irregular dimensions and special composition it is necessary to resort to special methods for slowing down the cooling after a normalizing treatment.

Here it is necessary to observe that the existence of a greater or lesser lag in the  $\gamma \rightarrow \alpha$  transformation depends in very large measure upon the effects reached by the same reheating. Other conditions being equal, the transformation will take place with more difficulty the greater the homogeneity produced by the preliminary reheating. It is easy to understand this point by referring to what was said in Chap. IX in connection with the relations between the dimensions and the position of the band of heterogeneous austenite and the temperature of the start of transformation. Its commencement is at a lower temperature the narrower the band of variant concentration and the further it lies toward the right in the fictitious binary diagrams.

tures, it is clear that the preliminary heat treatment is reduced to a reheating for homogeneity. The third of the series of phenomena indicated in Sec. 81 is missing in them as one determining the characteristics of the heat treatment.

Among the steels which belong to this category may be mentioned ternary steels high in manganese, containing from 9 to 14 per cent. of that element, or high in nickel, chromium, tungsten, and some other elements, besides many quaternary steels high in chromium and nickel, chromium and tungsten, etc.

These steels find but limited use as machinery steels today mostly because their excellent properties are not well known by manufacturers.

But there is no doubt that their applications will extend rapidly in that field, because their use provides very great resources and very important advantages, both in the manufacture of castings and in the manufacture of rolled or forged parts. I will have occasion further on to refer to some examples of high alloy castings.

85. The great majority of the steels employed at present for machine parts possess a transformation range at equilibrium well above the blue heat, nor do they exhibit very wide thermal hysteresis. Therefore, the selection of the preliminary heat treatment in each case and the exact determination of the particular conditions for its execution, involves taking precise account of the phenomena belonging to the third of the series indicated at the beginning of Sec. 81; namely, of the phenomena which occur during the  $\gamma \rightarrow \alpha$  transformation.

Upon this point the general directive criterion applied in the great majority of cases could not be simpler. It consists of an attempt to minimize the importance of the phenomena of liquation which we know always accompanies that transformation, causing fresh heterogeneity in the final mass, mainly of  $\alpha$  iron. The practical difficulties in each case arise only in determining the limits within which one may work in compatibility with the other necessities of steel manufacture, as well as with the necessity of avoiding various troubles inherent in the process.

Neglecting for the time being the practical difficulties just mentioned, and considering the problem from a more general and theoretical point of view, the general method to be followed to insure homogeneity by preventing liquation during transformation is easily found in the light of the considerations analyzed in Part II.

I have already pointed out in Sec. 82 that it was not practicable to apply a process analogous to quenching which is capable of suppressing liquation during solidification. Instead, it is almost always possible to apply a true quenching process during allotropic transformation, and it is easy to understand that when applied so as to produce fully its characteristic effects, quenching must necessarily suppress any liquation effects accompanying the transformation. In fact, it is known that

may be suppressed absolutely.

Therefore, it appears that the simplest method to prevent heterogeneity caused by allotropy consists in subjecting the steel to the most energetic quenching possible from a temperature above the initial point of the  $\gamma \rightarrow \alpha$  transformation. Other practical exigencies limit the selection of the operating conditions.

As a matter of fact it is necessary to resort to such so-called "homogeneity quenchings" to prevent variation in concentrations simply because it is not possible to resort to other means analogous to those which may be applied with good results to prevent or to minimize the heterogeneities produced by liquation during solidification.

To apply processes similar to those applicable for the last-mentioned purpose, one ought to reheat the metal to a temperature below that of the end of the transformation. But this is impracticable first of all because at such a temperature, in all the steels to which this study applies—particularly steels having very low temperatures of transformation—diffusion takes place at velocities so small that the effects cannot be felt in perceptible measure except when the reheating is prolonged for an extraordinary length of time. Under such conditions the method ceases to be applicable in manufacturing practice.

Almost all machine steels decompose during the  $\gamma \rightarrow \alpha$  transformation into substances which cannot all enter into solid solution with the resulting ferrite; that is to say, the composition lies within a field of insolubility (*lacuna di miscibilit *) see page 39. In practice, this latter fact has a far greater importance than the one mentioned previously. Hence, transformation is accompanied by the segregation of the two phases stable at temperatures below the critical. Now we know that the properties of the resultant metal depend largely upon the forms and special structures assumed by those phases, viz: saturated mixed crystals of  $\alpha$  iron, or ferrite, secondly free cementite, and thirdly the conglomerate of the two, or pearlite. Those structures are determined mainly by the conditions surrounding the cooling of the system through the transformation interval. Also we know from the fundamental principles of metallurgy that they cannot be modified by means of reheatings at temperatures below the critical, except in very small measure, when the metal possesses those forms which it ordinarily assumes upon slow cooling. Reheating such material causes such effects as are due to phenomena proceeding at an extremely slow rate; for instance, the coalescence or balling-up of the structural elements constituting pearlite, into what is known as spheroidal cementite.

We know, instead, that reheating to temperatures below the critical—a process more precisely indicated with the name of drawing—produces far more rapid and intense effects when applied to steels in which the

segregation of phases normally accompanying allotropic transformation has been partially or entirely prevented by quenching.

**86.** It is necessary to bear in mind some restrictions regarding the practical application of homogeneity quenchings.

Due to the fact that the first step in a homogeneity quenching is nothing but a process of reheating to high temperature, which likewise constitutes the first essential phase of any reheating for homogeneity, it is clear that the same observations and restrictions indicated in previous sections apply to both.

But other precautions and other restrictions essentially practical in character hold in the case of homogeneity quenchings.

As clearly indicated in what has been said in the previous sections, the maximum efficiency of a preliminary quenching will be obtained when the final phase, or quenching proper, is the most energetic possible. (This generally-used expression, "energetic," indicates that the steel has been cooled with the greatest possible velocity starting from the highest possible temperature.) This statement is on the basis that the conditions under which the first phase of the process has occurred may be disregarded. They have already been briefly investigated in the case of simple preliminary reheating. Now the indispensable practical restrictions to the use of a too energetic quenching depend essentially upon the fact that in industry a piece of steel cannot be quickly cooled in a uniform way throughout the entire mass of the metal, especially if intricate in form or of large dimensions. In rapid cooling the metal is subject to certain temporary displacements, due to the different thermal contraction of the various parts of the piece, at each instant, or to permanent ones, due to the different degree of transformation of steel in the various parts cooled at different velocities, and therefore of permanent contraction or expansion. Such displacements in turn give place to so-called "internal stresses" which may reach such an intensity as to break the piece.

It is clear that these phenomena have nothing to do with the inquiry under prosecution, therefore it would be out of place to investigate them at any length. It is enough to have indicated them in order to establish the fact that, in practice, it is not always possible to make the final cooling at a velocity as high as would be required to derive the utmost benefit from the homogeneity quenching.

**87.** In practice it is often possible to eliminate, or at least to attenuate, many difficulties in the heat treatment of large or especially intricate pieces, by inverting the terms of the problem, so to speak. The minimum cooling velocity necessary to obtain and maintain a certain degree of homogeneity is not an absolute value but depends upon the

of determining the most suitable quenching velocity for a given piece, to obtain a required residual homogeneity, one may try to solve the problem of finding a steel of such a composition which gives the same residual homogeneity after being subjected to a determined heat treatment. Having discovered such a composition, the quenching may be accomplished at the maximum velocity compatible with the dimensions, form, and nature of the piece under treatment.

Such a problem appears constantly in the metallurgy of special steels, and it may be said that all the steel plants specializing in this field, solve the problem every day, but more or less approximately. Due to the fact that such a problem is very complicated in practice, it very seldom happens that it occurs in the explicit form under which I have thought it useful to present it.

At any rate, from what has been said in the previous sections, it is clear that the solution of the question is to be sought in the employment of steels with a low transformation point. The transition velocity for such alloys is in general the smaller, the lower the temperature at which it occurs. The right selection of one of those steels allows the operator to obtain a given degree of homogeneity with a heat treatment which does not include any quenching whose "energy" (consequent to the heat from which the metal is quenched and to the velocity of cooling) exceeds a predetermined limit.

One may arrange steels of varying composition into a series whose properties show a gradual change from those of the plain carbon steels to those of alloy steels. The lower limit will be a metal which can be normalized to obtain the highest uniformity only when the treatments end with an energy equivalent to quenching. The upper limit will be the special steels in which the obtainable results may be had by a simple reheating followed by a cooling. All gradations of cooling practice, such as quenching in brine, in water, in oil, or in molten lead; cooling in air blast, in still air, in ashes or in furnace, are used nowadays in steel making when employing these steels, whose varieties are extremely numerous.

It may be noted here that steels with low transformation point may have their critical temperatures lowered in direct proportion to the degree of homogeneity reached in the  $\gamma$  solid solution. This may be considered according to the conclusion reached in the preceding chapters, that the beginning of the  $\gamma \rightarrow \alpha$  transition is the higher the wider the range of variant concentration.

Generally, no necessary connection exists between the most suitable temperature for homogeneity reheating and the correct temperature for the subsequent quenching, properly so-called. In fact, the first temperature is limited only by a minimum value below which the diffusion of carbon takes place at a velocity too slow to cause effects which can be utilized.



taking place within that interval.

The independence of these two temperatures is a fact of great practical importance, worthy of the greatest attention in the study of normalizing. Especially is this true when dealing with special steel of low transformation points. For the latter, only the suitable selection of two different temperatures may obtain the best results: namely, one for the reheating for homogeneity, and the other for subsequent quenching.

In other words, it is always necessary to bear in mind that a homogeneity quenching can never consist merely in a "quenching," in the narrow meaning of the word, but its effect on the properties of a steel always results from the sum of the diffusion during annealing and the subsequent checking or attenuation of the liquation inherent in the  $\gamma \rightarrow \alpha$  transformation, by means of a sufficiently rapid cooling.

88. It would be very easy to follow the gradual modifications that the equilibrium diagram of a given steel will undergo step by step with progressively increasing the cooling velocity. As was seen previously—in particular in Sec. 53 to 75—a more rapid passage through the  $\gamma \rightarrow \alpha$  transformation range results in a more incomplete attainment of the conditions of equilibrium in the mobile system. In other words, the heat treatment to which the steel is subjected gradually loses the characteristics of a true reheating and assumes those of a quenching proper.

These considerations are not analyzed here, because such a development would consist of nothing else but a repetition of what has been said before regarding the perturbations which are caused by each variation of the cooling velocity in the principal lines of the diagrams reproduced in Fig. 12, 13, 14 and 17.

89. In the great majority of cases normalizing is practiced by taking essential account of the principal aim, *i.e.*, to reduce segregation normally accompanying the  $\gamma \rightarrow \alpha$  transformation to the minimum value practically obtainable, without subordinating in any way the selection of the best means of reaching this purpose to the special physical and mechanical characteristics which the metal should eventually possess.

In general, it follows that normalizing (which for this very reason is called a "preliminary heat treatment") is followed by others more or less complicated, called "final heat treatments," which have as their essential purpose the adjustment of the metal to the special and oftentimes very narrow limits of mechanical properties often specified as necessary for its intended employment.

From what I have said several times, the analysis of the conditions under which these "final" or "quality" heat treatments must take place, would go beyond the limits of this study. However, it is timely to point

must not obliterate totally or in too substantial amount the useful effects obtained by normalizing.

In general, the result now indicated is almost always sought by operating in such a way that the final heat treatment does not involve reheating the steel to a temperature above the  $A_c$  transformation points. The subsequent cooling can no longer be selected at will, since its velocity is determined by the final mechanical properties of the piece. If, then, at this point, the range where the  $\alpha \rightarrow \gamma$  transition proceeds is entered, the reverse reaction on leisurely cooling may allow sufficient segregation to neutralize the effects of the previous equalization. In the practical application of this criterion, wide use is made of the phenomena of thermal hysteresis. It causes the  $\alpha \rightarrow \gamma$  transformation on heating to occur at a far higher temperature than on cooling, and so largely widens the range of temperatures which may be utilized in executing the final heat treatment under the best conditions.

90. Reheating to a temperature below the critical can be almost always utilized in practice in the simplest of all "quality" heat treatments. In this case there is no other purpose than more or less completely converting the metastable system to the stable one. In all but lowest carbon steels, a homogeneity quenching results in a complex, usually constituted mainly of austenite and martensite, but sometimes mixed with troostite. Reheating such material simply endeavors to produce the required mechanical properties in the finished steel by maintaining the temperature as low as possible and yet allow a determined amount of segregation in the stable phases, constituting troostite, sorbite, and pearlite. Evidently in such a case it is always a prime endeavor to avoid going above the transformation point on reheating. In other words, the final treatment is restricted to a true "drawing" operation instead of pushing the temperature up to the limits of a true "annealing." Rather than get a rapid rearrangement at high heat, it is then better to get the same result by prolonging the length of the heating necessary to impart to the steel the desired mechanical properties.

In other cases, such a restriction may not be imposed upon the temperatures to be reached in the final operation. As an example, recollect what happens every time the final heat treatment must include a "hardening." Essentially, hardening imparts a determined structure to the solid solution constituting the steel after quenching, corresponding, generally, to a given ratio between the first products of the transformation of austenite; *i.e.*, martensite and troostite. For the same reasons already given it is always the endeavor to limit temperature of heating as closely as possible and to maintain the metal at this "lowest maximum" temperature for the shortest possible length of time.

The realization of the conditions now mentioned becomes very difficult in many instances on account of the nature of the steel employed, the form of the pieces to be treated, and, finally, on account of different composition of the metal in various parts of the same piece. A piece which is thick in one section while thin in another, is delicate to handle since it may be impossible to reach determined heating or cooling velocities in some parts of them without causing those velocities to be far different in others. Variation in chemical composition may be the result of design, as happens, for instance, in partially case-hardened steels. On the other hand, a piece may be formed by various parts joined together in any way purposely, by welding or otherwise, or accidentally, as happens in raw steel which during solidification has undergone an intense segregation called "great liquation."<sup>1</sup> Discontinuities of this sort greatly increases the difficulties inherent to the selection of a final heat treatment rationally coordinated with the preliminary treatment. A practical solution of these problems often necessitates the most complicated programs.

91. In making a complete heat treatment, it is possible only in exceptional cases to disregard the considerations mentioned in the two previous sections. This happens when it is possible to select a preliminary heat treatment which, although being sufficiently efficacious as such, is also capable of imparting to the steel those special physical properties fitting it for use, alone and without any further treatment. This occurs very seldom. It may be due to an accidental coincidence between the most suitable conditions for the preliminary and the final treatment. Or less often it may be due to the less severe specifications which the metal must fulfill. In the last case, it sometimes actually happens that the wider tolerance allowable both for temperature and velocity of cooling when executing each of the two treatments, may cause an overlap of determined regions in the two fields. In that case it is sometimes possible to obtain results sufficient for limited requirements with only one treatment, by working within the limits of that common region.

An accidental coincidence of optimum heat treatments is the most rare of the two cases just above indicated. It occurs almost exclusively for very mild steels, in which the physical properties of the mixture of metastable constituents resulting directly from even the most energetic preliminary quenching corresponds quite well to the requirements which these steels must ordinarily satisfy.

However, it is necessary to bear in mind that, even in this most favorable case omission of the final drawing cannot be suggested except for a possible saving in expense. No matter how mild or low in carbon and manganese is the steel, its use after full quenching never insures the

<sup>1</sup> See page 150, paragraph 111-D.

guaranteed reliability confidently obtained when a steel a little "harder" in chemical composition is used after being suitably drawn following a homogeneity quenching. This is due mainly to the fact that drawing, when suitably done, is the only means for efficiently eliminating internal stresses caused by quenching, especially if energetic, as must be a homogeneity quenching for very mild steels with high critical ranges. These stresses may reach dangerous amounts especially in pieces of large dimensions and of intricate, unsymmetrical forms, at times even causing spontaneous fracture.

An insufficient or neglected drawing may often cause other troubles which, though less dangerous than breakage, represent a heavy loss. For this reason, false ideas of economy may result in net waste. As an example may be mentioned pieces of complicated forms, such as a 6-cylinder crank shaft, which must be machined with great precision. If these pieces are subjected to an insufficient annealing after the preliminary quenching they sometimes retain internal stresses capable of warping them so badly during the finishing that they are rendered useless. Such machine work cuts away layers of metal which formerly partially balanced those tensions.

92. An overlap in the preliminary and final heat treatments may for instance, occur when a steel is used for a certain purpose requiring much less than its full possibilities. Thus, in working a steel belonging to a type harder than would be required, one could replace the true preliminary quenching by an intermediate treatment lying somewhere between a quenching and a simple annealing.

The nature of such so-called "incomplete quenching" has been explained before at sufficient length when discussing special steels of low transformation-point. Therefore, it is superfluous to return to that argument. Considering the process from the special point of view which we are now particularly concerned with, namely, that of the final physical properties of the steel, it must be added that other things being equal the greater difference between the results obtainable by an energetic preliminary quenching followed by drawing and the results obtainable by an "incomplete quenching" (more or less similar to a reheating) consists in the fact that in the second case the metal shows an elastic limit very much lower than in the first, yet without an equal difference between the values for tensile strength. In other words, the double treatment produces a much higher value of the so-called "elastic ratio" of the metal (elastic limit divided by the ultimate strength, a ratio almost always greater than one-half).

When the uses to which the steel will be put do not require a high elastic limit, or when it is possible to use a harder steel than necessary

be sometimes convenient to employ the process of "incomplete preliminary quenching" just described.

A comparison will be given further on, based upon experimental figures, of the differences between the physical properties and the structures obtained by each method. As far as the structures are concerned, it is even now easy to account for these differences, if one bears in mind the considerations developed in Part II. The relation between the degree of homogeneity and its influence upon the course of the phenomena of segregation accompanying the  $\gamma \rightarrow \alpha$  transformation were discussed there at length.

Perhaps it is superfluous to point out that, in the industrial application of the process of "incomplete quenching" just described an accurate study is necessary for each individual case to determine the more suitable "limitations" to the process of "normal quenching" in order to produce an "incomplete quenching." It is known that these limitations may consist essentially of taking a longer time to cross the transformation ranges; or else in interrupting the rapid cooling of the quenching proper at a temperature below the initial point of the  $\gamma \rightarrow \alpha$  transformation, from which point follows a much slower cooling to room temperature.

## CHAPTER XIII

### PHENOMENA WHICH MAY DISTURB THE COURSE AND MODIFY THE EFFECTS OF PRELIMINARY HEAT TREATMENTS

93. Before closing the general discussion forming Part III, it may be said that the action of heat treatments is liable to many disturbances due to a great number of different causes, some of which produce effects of the greatest practical importance.

It is not possible to make a complete study of these causes and effects in these pages. Two groups of facts may be mentioned as examples, selecting them from among those which occur more frequently in steel making and which have the greater practical importance. They have profound effects upon heat treatment processes; some concrete examples of their ultimate effects will be reserved for later mention.

The first of these groups includes occurrences during hot-work, such as rolling, pressing, forging, or drop forging the steel, evidently operations entirely normal to ordinary metallurgical practice. Yet they may cause important changes in the course of the phenomena underlying preliminary heat treatments and result in strong modifications in the properties which may be obtained in the same steels by means of a given heat treatment.

On the other hand, the second group includes phenomena which ought, theoretically, to be considered as being accidental. They are connected with the presence of solid non-metallic inclusions or "sonims"<sup>1</sup> formed in the various phases of melting, refining, pouring, teeming and crystallization of steel.

It has been remarked that the phenomena due to the presence of solid non-metallic inclusions in steel ought to be considered as accidental. Sonims certainly cannot be considered as normal constituents. On the contrary, they are extraneous bodies held in a metallic mass which remained too short a time in a sufficiently fluid state to allow such inclusions to separate from the mass by reason of difference of density. But I added that this fact can be stated only "theoretically" because, in reality, it may be said that no steel exists if made under normal conditions of manufacturing practice which does not contain solid non-metallic inclusions in larger or smaller proportions.

<sup>1</sup> We shall use the accepted word "sonims" throughout in place of solid non-

Some numerical data concerning special cases will be given later showing how the presence of slag inclusions may strongly alter the effects of preliminary heat treatments.

94. It may now be indicated how some of the general characteristics of hot-work place it within the sphere of the processes studied in the previous chapters. Later, a few particular cases will be cited to illustrate its technical effects.

It would be easy to demonstrate on the basis of actual experimental data the fact dwelt upon later that the characteristic effects of hot-work are not due to compression, except perhaps in a very small measure which certainly may be entirely ignored in practice. There can no longer be a doubt about the correctness of such a statement although it is openly in disagreement with ideas generally accepted by the so-called "practical" men.

By mere exclusion the effects produced by such mechanical operations are essentially due to the *deformations* which the steel suffers, a conclusion fully confirmed by experience.

It is necessary to distinguish two series of effects caused by such deformations. In practice, the phenomena of both series always intervene contemporaneously in determining the net effect of rolling, forging, and the like. While one series has hardly any noticeable effects upon the course and results of a normalizing treatment, the other series deeply modifies the issue.

Phenomena of a nature similar to those in Sec. 82, belong to the first series, namely, those connected with modifications of crystallographic structure, properly speaking. Numerous and very interesting observations exist regarding this class and permit the mechanism to be clearly and definitely explained. They show characteristics entirely analogous to those already thoroughly studied by the science of crystallography. Movement of the hot metal under outside forces is essentially a "sliding" or "reciprocal slipping" of crystalline elements or portions of such elements along definite surfaces, and of more or less well developed "twinning." The trend of these phenomena in the course of hot-work and the conditions of temperature under which they take place, exercise the greatest practical influence, causing intense modifications of the physical properties of the resulting piece. This statement holds regardless of the direction in which the article is tested, whether with or across the direction of forging. Therefore it is evident how great is the practical importance of an accurate study of these circumstances.

However, as has already been indicated, experience shows and simple theory readily supports the experimental conclusion that the mechanism of slip does not perceptibly modify the operation of diffusion and other fundamentals upon which normalizing practice is based. It is, therefore,

although it would be indispensable as a basis of any study of mechanical hot-work.

Conditions are different for phenomena of the second series. Deformations which steel suffers under the action of hot-work, produce large reductions of one or more of the dimensions of the primary and secondary crystalline elements. It is clear that this last fact must exercise a very strong action upon phenomena of diffusion, which indeed constitutes the main basis of preliminary heat treatment.

From the discussion given in the first two parts of our study, and other conditions being equal, it is easily understood that a large deformation in any direction will correspondingly reduce the distances between the points corresponding to the maximum and minimum values of the concentrations of the various elements in solid solution in iron, and must lower the time necessary for a given homogeneity treatment to cause a determined effect.

These expectations are fully confirmed, and we shall later see some examples which prove, by experience, that a given normalization is far more efficacious when applied to a suitably forged or rolled steel than it is when applied to the same steel in the raw state.

95. Apropos of the phenomena which have been discussed in the previous paragraph, one may point out that the observations regarding the crystallization of steel, indicated in Sec. 81, also hold true regarding the practical application of the processes based upon the phenomena accompanying hot-working. It seldom happens except in very rare and commendable cases, that the study and the practical application of the processes aimed to suitably regulate the course of those phenomena, are a part of the duties entrusted to the man who has charge of studying and applying all the other processes constituting the ulterior steps in the heat treatment. For instance, in nearly all steel plants the superintendent of the rolling mill or forge shop has nothing to do with the heat treatment, and *vice versa*.

This place is hardly suitable to discuss instances and methods whereby the lines of authority could be advantageously modified. I only want to point out the fact in order to emphasize the necessity of perfect coordination in the plant as ordinarily organized between the operating methods followed in the mill and in the heat treatment department.

To be convinced of this it is only necessary to remember the before mentioned facts, that the selection of the most suitable heat treatment is very strongly influenced not only by the special conditions under which the steel has been made and cast, but also the hot-work to which it has been subjected. From this special point of view in particular, the temperature at which mechanical working has been done has a preponderant practical importance beyond all the other conditions. Merely as an



or even simply finished at too low a temperature, it gives the piece of steel certain peculiar properties due to cold-work.<sup>1</sup> An overstrained steel cannot be reheated above a determined limit, generally very low, without seriously "coarsening" the grain (see Sec. 82), a structural occurrence always accompanied by a great increase in fragility. The correct heat treatment of such material must be subordinated to special limitations of time and temperature whose determination is very delicate and difficult. Neither can it be made rightly except by a person possessing all the data concerning the conditions under which the mechanical work has taken place.

Phenomena of plastic deformation assume great importance especially when working certain groups of special steels. Mechanical properties must be imparted to them which probably can be reached only by means of mechanical and heat treatments executed under rigorously determined conditions within extremely restricted tolerances.

For reasons which have been repeatedly outlined, it is not possible to deal, in detail, with these phenomena here. They have been mentioned only as an example of facts which may modify the normal train of events underlying normalizing practice.

Merely as a matter of information, I wish to add that some steel plants, particularly those manufacturing special steels, have sometimes eliminated the evident inconveniences due to separate responsibility for the technical operations comprising the various steps of steel-making and working, by creating a kind of metallurgical supervisor-general, to observe all details of the manufacturing processes and report directly to the technical manager of the plant.

96. The second group of facts which have been mentioned in Sec. 93 as seriously modifying the trend of events analyzed in the first two parts of this study are due to the presence of non-metallic solid bodies included in the steel.

I intend to give further on some actual examples of these phenomena, as well as of those of the previous groups, designed to give a sufficiently exact idea of their real practical importance. However, it is now advisable to record some general observations concerning the particular characteristics by which they are more directly connected with the subject matter of the previous chapters. Inasmuch as the topic is of the greatest practical importance but usually studied very cursorily in treatises on metallurgy and metallography, especially from the point of view which more directly connects it to the phenomena which we are studying, it will not seem amiss to dwell here a little longer upon it.

A large variety of non-metallic solid inclusions are found in steel

<sup>1</sup> Following Howe, "plastic deformation" in the cold is identical with "cold-work" (French "ecrouissage," German "Kaltrecken," Italian "incrudimento") and "overstrain."

made by various processes now in operation. It is also known that the practical consequences resulting in a given steel from the presence of too large a quantity of any kind of nonmetallic solid inclusions may be very severe. Yet even extremely small quantities of inclusions are fatal if they belong to certain determined types.

In order that the conceptions of the nature and characteristics of the influence of certain non-metallic inclusions may not be too undetermined and vague it is necessary to explain first some notions concerning the nature, origin and the principal properties of the various types of those inclusions which are more frequently found.

In usual works' parlance, all visible non-metallic solid inclusions are usually indicated with the same generic designation of "slag inclusions." Indeed nothing prevents the adoption of such nomenclature provided it is clearly understood that the word "slag" shall not imply the notion that those inclusions are particles of that slag, properly so-called, which constantly forms upon the molten metallic bath in all steel-making processes. In fact, it now is known with certainty that many inclusions appearing in finished steel have nothing to do with the true slag whose formation is an integral and essential part in steel-making and refining. In order to avoid any misunderstanding, it is better, therefore, to adopt the expression "solid non-metallic inclusions" or "sonims."

This explained, let us briefly note the principal types of inclusions, differentiating the ones which exercise an appreciable action upon the course of the phenomena with which we are concerned. Then we shall see further on how it is possible to formulate some general hypothesis to explain the mechanism of these effects and to deduce from it useful indications regarding the way of counteracting them.

**97.** Accurate microscopic examination of all steels—especially if made upon perfectly polished samples before being etched, and observed under high enlargements—always discloses a more or less large quantity of solid non-metallic inclusions very small in dimension (of the order of a few hundredths of a millimeter) and of uniform and characteristic forms and colors. They are corpuscles of roundish form and grayish color, scattered in irregular manner in the various parts of the mass.

Figure 23 reproduces at 200 dia. an example of the more frequent appearance which these inclusions present. This micrograph refers to an unetched section cut from an ingot of mild carbon steel made by the acid open-hearth process.

Inclusions of the type illustrated are usually designated by the term "emulsified" inclusions. They occur in nearly all steels obtained by fusion in any of the commercial processes now used. The dimensions and the number of the grains is the only variation. We shall see soon that

as well as between those particles and the normal structural elements of steel, change with the conditions under which the steel has been melted, refined, poured, teemed and treated. We shall also see that these variations make their effects felt sometimes in very large measure upon the mechanical properties of the finished metal.

For this reason emulsified inclusions may be considered as those which exercise the greatest modifications upon the phenomena accompanying heat treatment, and, therefore, *indirectly* upon the *general* physical properties of steel. From this point of view their industrial effects differ decidedly from those produced by inclusions belonging to other types. Either because of their greater dimensions or difference in shape and constitution, these latter also exercise a *direct* action upon the properties

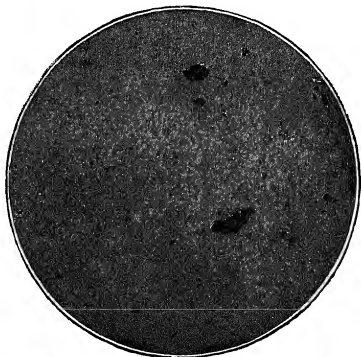


FIG. 23.—Emulsified inclusions in acid open-hearth mild steel.  $\times 200$ . Unetched.

of steel, sometimes even a great deal more intense than those produced by the emulsified slags. However, this action is permanent in the sense that it is not appreciably modified by heat treatments, and in addition has a localized influence. These last words should be understood in the sense that the solid non-metallic inclusions other than "emulsified inclusions" appear with a much smaller frequency (that is to say, the number of the particles is much smaller per unit of volume), and are ordinarily distributed in a much more irregular manner throughout the metallic mass forming isolated "groups" or "nodules." As a consequence it is always easy to detect localized modifications in mechanical properties of the metal around the single inclusions or groups of them. These

that when the metal is tested to destruction the fracture exhibits special characteristics and is located at the inclusion or groups of them.

98. Following the program which has already been indicated at the beginning, I should give a concrete idea of the specific characteristics of the modifications and alterations caused by emulsified inclusions, only after having tried to give a like concrete idea of the normal effects produced by the preliminary heat treatment.

However, inasmuch as I have already mentioned the external characteristics which distinguish emulsified inclusions from those of other types, I think it timely to briefly indicate the relative positions normally held by the individual particles of sonims and the ferrite, cementite, pearlite or other well-known structural elements in our steels. In this way, when



FIG. 24.—Medium steel with inclusions enclosed in ferrite.  $\times 80$ . Etched with picric acid.

the modifications of mechanical properties shall be exhibited by means of actual examples of a determined preliminary heat treatment upon steel containing emulsified inclusions, the modifications produced by such inclusions upon the normal structure of steels will also be much clearer.

Ordinary relations between particles of emulsified inclusions and the structural elements of steel may be easily seen in microscopical examination of the raw steel, just as it has been poured. Polish the sample and etch slightly with one of the usual reagents, such as an alcoholic solution of 5 per cent. picric acid, or a solution of nitric acid in amyl alcohol.

In the great majority of cases the emulsified inclusions are found

ing the direction of their eventual alignments so that said inclusions are not in contact with pearlite.

The two following figures show two characteristic examples of such structures. The first (Fig. 24) magnified 80 dia., reproduces the structure of a medium-hard carbon steel in which the emulsified inclusions appear isolated and surrounded by a nimbus or halo of ferrite.

Figure 25 reproduces at 80 dia. the structure of another medium hard carbon steel, in which the emulsified inclusions appear in orderly lines, and each string is entirely surrounded by ferrite.

It happens only in rare cases that raw steels which have not yet been subjected to any work or heat treatment do not show either of these

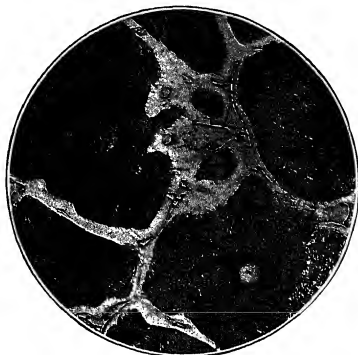


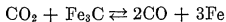
FIG. 25.—Emulsified inclusions in ranks directing disposition of ferrite veins.  $\times 80$ . Etched with picric acid.

characteristic arrangements. In such cases the emulsified inclusions are partly or entirely in contact with pearlite.

An example of such is shown in Fig. 26, which reproduces (magnification 120 dia.) the structure of a medium hard 2 per cent. nickel steel. The figure does not require any explanation.

A series of experiments<sup>1</sup> makes it practically certain that the occurrence of elements of ferrite surrounding the particles of slag or its independence from the position of such particles depends upon the state of oxidation in the system. Or, better, it depends upon the ratios between the oxygen pressure of the non-metallic particles and of the metal, respectively, and from the variations which this ratio suffers during crystallization and the subsequent cooling. When these variations are

Such oxidation is due to an excess of  $\text{CO}_2$  in the atmosphere of dissolved gases above the amount in equilibrium with CO and C in the average surrounding metal. By the well known reaction



excess  $\text{CO}_2$  shifts the reaction to the right, decarburizing cementite and producing a layer of ferrite around each element of the inclusions at a given instant of the process. This first layer acts during subsequent cooling of the system through the transformation interval  $\gamma \rightarrow \alpha$  as a crystallization germ attracting the further quantities of ferrite which separate during the critical range.



FIG. 26.—Medium 2 per cent. nickel steel containing emulsified inclusion.  $\times 120$ .

The reasoning which leads to this conclusion is based upon many observations of different nature which are not proper to develop here, but which are in part founded upon the study of the modifications which heat treatment produces in the relations between emulsified inclusions and ferrite.

It is clear that such reactions impress an orientation to the ferrite and pearlite which differs greatly from the characteristic "normal" orientation whose genesis we examined in the second part of this study, because an oxidized inclusion surrounds itself with a thin sheath of ferrite, which in turn accumulates much more ferrite by acting as a series of crystallization centers. It is now seen that the presence of emulsified inclusions has immensely altered both the normal segregation of ferrite

accompanying the  $\gamma \rightarrow \alpha$  transformation as we have studied it up to now, and the structures which are thereby produced.

99. Since "emulsified inclusions" are nearly always present in steel, they perhaps assume almost a "physiological" character, stretching this adjective to mean that their occurrence practically is a necessary consequence of normal steel-making processes.

On the contrary, one may say that the presence of non-metallic solid inclusions belonging to all the other types constitutes a phenomenon of a "pathological" character, meaning that such inclusions are formed and have been retained in the metal by an abnormal and defective course of one or more of the steel-making operations.

It is necessary to qualify these statements by remarking that the presence of the emulsified inclusions may also assume a pathological character in a steel too highly "oxidized" when they appear in thin laminations (Fig. 25) and when at the same time their effects upon the agglomeration and orientation of the ferrite are of exceptional intensity. Contrary to what is true for the other types of inclusions, the pathological character of emulsified sonims is indirect, inasmuch as it is due not to the inclusions *per se.*, but to their influence in the special disposition or positioning of the ferrite, which latter is of course a perfectly normal structural element of steel. In the greater majority of cases this effect is happily susceptible of strong modifications, and sometimes complete elimination, by means of suitable heat treatments.

100. Owing to the immense practical importance which the presence of non-metallic solids assumes in the study of the problems related with steel making, treatment and usage, and admitting the profound differences indicated in the previous paragraph between the behavior and the technical effects of various types of inclusions, it is clear that an exact knowledge of the nature and origin of such inclusions would have a very great industrial importance.

The problem has been already studied attentively, although perhaps not with the thoroughness which it deserves. It is now possible to describe precisely the nature and the origin of some types of inclusions, while of many other types it is not possible to say how they arrived nor what is their constitution.

True enough, emulsified inclusions fall within this *terra-incognita*, even though as we have seen and shall see still better they exercise the greatest effects upon the course and results of homogeneity heat treatments. It would be impossible to present here with sufficient brevity a

Therefore, I prefer merely to indicate briefly the hypothesis concerning the nature and the origin of the emulsified sonims which appears most plausible at the present state of our knowledge.

Opinions of various experimentors are extremely various concerning their chemical constitution. However, none is founded upon sure and direct experimental data such as could only be furnished by a chemical analysis of the material actually constituting the inclusions. It has not yet been possible to make such an analysis due to difficulties inherent to a mechanical separation from the surrounding metal in order to gather a sufficient quantity for testing. Chemical methods for the separation of inclusions from the metal are founded upon the employment of solvents for the iron. Such a procedure has not yet furnished reliable results, because all the reagents suggested cause more or less profound alterations in the material forming the inclusions.

It would take too long, as well as being out of place here, to list the conclusions reached by various experimentors on this subject based upon various methods of indirect investigation. It is enough to say that according to the opinion of many, emulsified inclusions are constituted exclusively or almost entirely by manganese sulphide or by mixtures of sulphides of manganese and of iron. According to others, these inclusions would be usually formed by oxysulphides of iron and of manganese, while there is another who thinks that they also contain silicates. Finally another opinion seems perhaps more likely, both because it conforms better with plausible hypotheses advanced to explain the origin of other types of inclusions, and because the more common emulsified inclusions show substantially the same properties in steels relatively high in sulphur content, as well as in those almost free from sulphur. According to this opinion, emulsified inclusions are essentially silicates in addition to sulphides, silicates of iron and of manganese, united to other silicates—as those of calcium, aluminum, magnesium, etc.—whose nature depends upon the kind of refractory materials with which the molten steel has come in contact. The proportionate amount of sulphides would vary with the average percentage of sulphur in the materials used in making the steel; but in general it would be somewhat low, so as to leave the preponderance of mass to the silicates.

In other words: the composition of the emulsified inclusions would be fundamentally similar to that of the slag proper which forms during the operations of steel melting and refining, and which appear floating upon the molten metallic bath at given periods of the process.

**101.** Regarding the origin of non-metallic inclusions, it is certain that it varies a great deal for the various type of inclusions.

The more persistent, and to us more interesting, emulsified sonims are attributed by a very well founded hypothesis to an origin in phenomena



former cannot mix. It is from this analogy that the expression emulsified inclusions has originated.

Inasmuch as it is impossible to enter upon a detailed discussion of the reasons why the hypothesis above indicated appears credible, we will only outline its content.

During steel-making the formation of slag constantly accompanies the chemical reactions occurring along with fusion and refining, whatever the manufacturing process may be. These molten non-metallic substances forming the slag proper exist largely in the form of globules of different dimensions in suspension in the liquid steel, when it is more or less agitated by the energetic reactions which take place in it accompanied by evolution of gas. As soon as this period of agitation, which in certain measure always appears in all "fusion" processes, comes to an end, it is succeeded by a quiet period during which the final deoxidation reactions take place in the metallic bath upon the addition of deoxidizing alloys, mainly ferro-manganese, ferro-silicon, or other alloys having silicon and manganese as basic elements. During these relatively quiet periods, all suspended slag globules tend to separate on account of a difference in density, a process analogous to that taking place when settling a turbid liquid. However in our case perhaps it is more exact to make a comparison with an oily emulsion left in repose, when the oily drops gather little by little in groups, coalescing into larger droplets by the action of surface tension. These large drops ascend toward the surface of the denser supporting liquid, at a more rapid rate the larger their dimensions.

Spring was one of the first to study this subject accurately. He indicated the causes of clarification of such mixtures under the general expression *floculations des milieux troubles* (flocculation of turbid liquids). While flocculation is a pronounced reason, it is reinforced by the viscosity and some other properties of the dispersion medium very difficult to evaluate with the usual physical and chemical criteria. One needs only remember, in this regard, that the velocity of sedimentation is enormously increased in suspensions and watery emulsions by the addition of very small quantities of some electrolites. It is also true that the entity and the nature of the phenomenon of sedimentation vary within wide limits with the *nature* of the added electrolite.

**102.** Now consider the mass of steel as an emulsion immediately after the period of agitation ceases; namely, after the bath ceases to "boil" in the Siemens-Martin refining process, or after the Bessemer "blow."

If we suppose that the process of separation of the suspended slag starts at this point, due to difference of density, and proceeds according to laws similar to those which regulate the corresponding separation occurring in aqueous emulsions, we may expect the following principal occurrences:

(a) Drops in suspension, of size larger than a certain "critical dimension" will ascend with relatively high velocity toward the surface of the bath, there to unite with the floating layer of molten slag.

(b) The same thing will happen for the drops resulting from the agglomeration of the smaller, step by step as their increasing dimension will have reached the critical.

(c) Suspended droplets which have dimensions equal to or smaller than the above indicated limit will rise in the bath with velocities extraordinarily smaller than that reached by the larger drops mentioned in (a) and (b).

Consequently, after a certain period of time, the undisturbed bath will no longer contain inclusions of molten slag, except those which are in the form of minute drops whose dimensions are equal to or smaller than the "critical"—a fact very well established for aqueous suspensions and emulsions.

Under normal conditions of temperature and for steels of ordinary composition, the time necessary for substantially complete separation of the inclusions indicated in (a) and (b) is relatively short and of such an order as to render it possible in practice to maintain the molten metallic bath quietly so as to eliminate those inclusions.

However, the velocity with which the particles of the dimensions specified in (c) rise in the molten metallic mass is so far below that of the others, that it would be practically impossible to keep the melt in quietude during a time long enough to allow their separation. Therefore, these remain in the metal and constitute those particles which we have named emulsified inclusions appearing in solidified steel.

**103.** The hypothesis briefly outlined in the two previous sections corresponds well to characteristics shown by the principal types of non-metallic solid inclusions. It explains the normal presence of the emulsified inclusions in steel and their constant characters. It also explains the fact that the presence of other types has the characteristics of an accidental and abnormal phenomenon. A further similarity to liquid suspension lies in the fact, well known to the steel melter, that with each variation in the quality<sup>1</sup> of the finished steel, the nature, state and number of the non-metallic solid inclusions also varies within very wide limits, other conditions being equal.

This last fact is only in part due to the different "viscosity" which characterizes various steels under normal conditions of melting and pouring. In fact, in many cases it certainly appears that the viscosity of the metal is not competent to explain the extreme variation in numbers and dimensions of the inclusions. Therefore, it is necessary to admit the

<sup>1</sup> By the term "quality" is meant not only the chemical composition determinable

intervention of other causes which might be looked for in phenomena similar to those caused by electrolytes upon colloidal suspensions.

This is not the place to discuss the probable nature and the characteristics of these last phenomena as they apply to steel-making. Besides, very little accurate information is known on this subject.

**104.** The considerations briefly condensed in the previous sections assign a character and a definite field of existence only to the non-metallic inclusions originating from the slag proper. Of these, it has already been said that only the inclusions whose dimensions are less than a certain definite limit (emulsified inclusions properly so-called) show peculiarities which might almost induce one to consider them as normal structural elements of the steels made by modern commercial processes. Therefore,



FIG. 27.—Large particle of slag in steel.  $\times 50$ .

it is necessary to keep account of these alone in a study of the general properties of steel. And inasmuch as their effects upon the object of our study may sometimes assume a great material importance since they are capable of modifying the general properties of steel obtained through normalizing treatments, I will mention, later on, some example of those effects. Instead, all other sonims without distinction must be excluded from the number of substances which normally modify the general properties of steel and therefore from the number which is necessary to take into account in a general study of the preliminary heat treatment of steels.

**105.** Merely as a matter of information, it may be timely to add some examples of those other non-metallic inclusions which simply represent accidental defects in the steel

Bearing in mind the hypothesis formulated in the previous sections, the first to be taken into consideration among the inclusions of this category are those originating from the metallurgical slag but which,



FIG. 28.—Slag inclusion showing two constituents.  $\times 100$ .



FIG. 29.—Globular slag inclusion.  $\times 100$ .

normally, should separate before the solidification of the metal—namely the inclusions which in Sec. 102 have been classified under the letters (*a*) and (*b*).

These inclusions sometimes reach large dimensions. An example is reproduced in Fig. 27, enlarged 50 dia. Often they show distinctly the two characteristic and well known constituents of the slag from



FIG. 30.—Serpentine lamination of slag.  $\times 100$ .



FIG. 31.—Slag inclusion of unknown origin.  $\times 100$ .

whence they come. A very evident example of this fact is given by the inclusion reproduced in Fig. 28, enlarged 100 dia.

Frequently inclusions of this kind are found in the slag from the

or drops, a fact which confirms the explanation given above of their origin. An example is given in Fig. 29, enlarged 100 dia.



FIG. 32.—Particles of refractory caught up by molten steel.  $\times 100$ .



FIG. 33.—Particle of refractory caught up by molten steel.  $\times 100$ .

It also often happens, especially in steels highly "oxidized" in melting, that the large inclusions assume the form of long, thin laminations, twisted as serpentine filaments in the section examined

enlarged 100 dia.

In order not to dwell too much upon this subject, which presents



FIG. 34.—Group of inclusions in heat-treated forging.  $\times 70$ . Etched with 5 per cent. picric acid in alcohol.

but a negative interest to normalizing practice, Fig. 31, 32 and 33 at 100 dia. are given to show three examples of the appearances which may be presented by inclusions having neither the characteristics of the emulsified inclusions, nor those of slag particles proper. The origin of such inclusions may be widely different, nor do we need to investigate it here. In passing it may however be said that many reasons lend to the belief that inclusions of the type reproduced in Fig. 32 and 33 are only particles of some of the refractory materials with which the molten steel comes in contact during its manufacture, either in the furnace, in the spout or in the ladle, or, in some cases, after its teeming from the ladle, as happens for instance, when the ingots are bottom-poured through runners or have refractory tops or risers.

106. In order to finish these, somewhat random observations some examples of those "local" effects may be added which, as has been said, are caused by other than emulsified inclusions. As has previously been mentioned, the latter are the only ones which may be considered as possessing the property of altering the *general characteristics* of steel. In this way the effects produced will be more precisely defined, even before presentation of some examples which are more directly related to the

specific results of preliminary heat treatments.

The *local* effects produced by sonims other than emulsified inclusions show a great number of different aspects in steel-making practice. Such effects may be chemical in character. For instance, excessive corrosion

may be localized around the individual inclusions. On the other hand, the effects may be mechanical. Some of the latter, (for instance, the abrasions caused when the inclusions happen to occur upon a surface subjected to friction) show characters so different from those produced by emulsified inclusions in relation to the results of heat treatment that really it is not necessary to illustrate with examples.

Therefore, there are left the effects caused by non-emulsified inclusions upon the behavior of steel when subjected to the ordinary physical tests usually specified. Of these extraordinarily various examples I will give only two.

The first is afforded by a medium hard 2-per cent. nickel steel containing non-metallic inclusions of large dimension united in groups which forged down into elongated shapes looking like stripes in the micrograph. Figure 34, enlarged 70 dia. after etching with 5 per cent. alcoholic solution of picric acid, reproduces the appearance of such a group of inclusions taken after a very accurate and efficacious heat treatment has eliminated (by a process which we shall see further on in regard to the emulsified inclusions) the large strip of ferrite which had collected around the group in the forged steel, due to the phenomena seen in Sec. 98. Now, in spite of the efficiency of the homogeneity heat treatment, proved by the elimination of this ferrite accumulation, the inclusions continue to exercise very intense local effects easily detected in the tension tests made upon the metal by means of transverse test-bars taken with their longitudinal axis perpendicular to the direction of the elongation produced by forging. Some tension tests made upon a certain number of test-pieces whose fractures did not present any anomaly, gave satisfactory results varying between the following limits:

Tensile strength.....	92,000 to 101,000
Elastic limit.....	64,000 to 70,000
Elongation.....	21 to 18 per cent.
Reduction of area.....	43 to 35 per cent.

Other test-pieces gave very poor results, extremely low tensile strengths, their elastic limits could not be ascertained and they showed very small elongations and contractions almost zero.

The fractures of these last named test-pieces clearly showed local defects of the same type as that reproduced at natural size in Fig. 35. It is easy to find under the microscope that the heterogeneous white strip which shows upon the fracture is nothing else but the location of a series of lamellar inclusions. Due to their fragility and to their insufficient adhesion to the metal, they have caused the well known effects of a sharp notch made in the metal, acting as "primer" and provoking the



The test-piece reproduced in Fig. 35 gave the following results:

Tensile strength.....	47,000 lb. per square inch
Elastic limit.....	Could not be determined
Elongation.....	1.8 per cent.
Reduction of area.....	<i>nil</i>

When testing such steel parallel to the direction of working, the presence of inclusions gathered in isolated strips such as the ones just illustrated causes great deal less perceptible effects than those which have been mentioned for the transverse test-pieces.

A second example is afforded by a medium-hard carbon steel made in the acid open-hearth and forged from 7-ton ingots to round bars 200 mm. in diameter. The test-pieces taken from all parts of the bars gave excel-

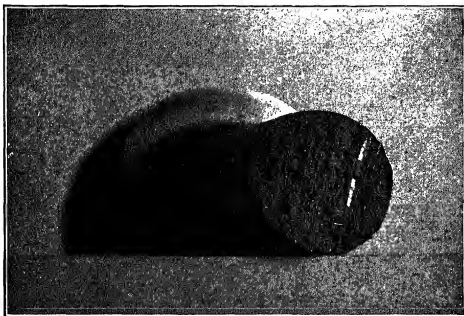


FIG. 35.—Streaked fracture due to lamellar inclusions. Natural size.

lent results, except those from near the *bottom* of the ingot. These gave very low tensile strengths, the elongations and contractions were almost zero, even when the tests were made upon longitudinal test-pieces. The fractures of the latter showed appearances similar to the one reproduced at normal size in Fig. 36. Microscopic examination of these test-pieces demonstrated that they contained numerous non-metallic inclusions of the same appearance as that reproduced in Fig. 27. A more detailed study of the foreign bodies and a chemical analysis of a small quantity which could be separated mechanically from the steel, showed them to be constituted of acid slag formed during working the heat.

**107.** Finally, it is necessary to call attention to a group of phenomena which sometimes occur as an effect of the same operations to which the steel is subjected during heat treatment, although they are not of a nature

normalizing treatments are based. Due to the fact that their effects accrue to those directly due to heat treatment, they often cause very noticeable modifications in the final physical properties characterizing the various parts of the steel object treated.

The phenomena which are to be mentioned are the following: When an object somewhat complicated in form is quenched, the cooling velocities of its various parts are unequal; they reach the minimum values in the

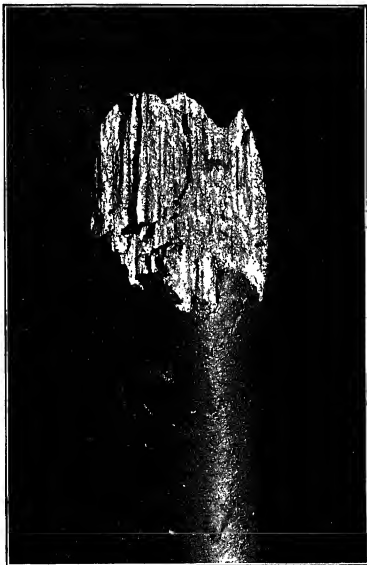


FIG. 36.—Woody fracture due to inclusions of furnace slag. Natural size.

thickest parts and the maximum values in the thinnest. Such differences of quenching speed sometimes cause very important internal stresses in the steel, which in turn, produce deformations, or warping.

Slipping and twinning are produced in the crystalline elements in certain parts of the piece, under the action of these deformations. These crystalline movements are analogous to those which have been already indicated in Sec. 94, when discussing the effects of hot-work.

The effects of these crystallographic phenomena sometimes are simply additive to those caused by heat treatment proper; but, in other cases, profound modifications of the results directly due to the heat treatment magnify this simple addition effect. For instance, this last statement holds when the effects of the deformations already mentioned continue to be felt after the piece has reached, during either rapid or slow quenching, temperatures very much below those which mark the limit called ordinarily, but erroneously, "limit of plasticity." In these cases, a strain-hardening occurs analogous to that when steel is subjected to distortion at a temperature too low for correct working. The characteristics, the consequences, and the practical value of those processes have been also mentioned in Sec. 95, but it is not possible to dwell upon their study for the same reason then pointed out. It is sufficient to keep in mind that the practical importance of the phenomena is very great. Therefore it is necessary to take them into very strict account in the industrial application of heat treatment processes although, as we have seen, those phenomena are not related with the fundamental ones under study. They constitute rather one of the more delicate sides of the technical problems inherent to the preliminary heat treatment of steels.

## PART IV

### PRELIMINARY HEAT TREATMENT OF STEEL CASTINGS

#### CHAPTER XIV

#### GENERAL MORPHOLOGICAL RELATIONSHIPS BETWEEN THE SECONDARY AND THE PRIMARY CRYSTALLINE SYSTEMS

108. The general characters assumed in practice by those complex phenomena comprising the foundation of normalizing technique were briefly indicated in the third part of this study. This was done mainly with the view of demonstrating that the effects of the fundamental phenomena studied in Chapters 1 and 2 are almost always modified in practice by the coincidence of other phenomena whose effects are sometimes simply additive, while at other times they give rise to profound modifications of the course of the fundamental phenomena. In order to complete the development of our program all that is now needed is a certain number of examples designed to show the special technical aspects which characterize the more typical cases occurring in practice and to give some numerical data to define the quantitative values of the results which can be obtained by the application of preliminary heat treatments.

We have already seen in Sec. 94 and 95 the immense practical importance of the effects of hot-work upon the proper development of the phenomena accompanying heat treatment. The application of such treatments must therefore be conducted according to adjustable rules, depending upon whether the steel to be treated is in its cast condition (raw) or whether it has already undergone forging, rolling, or other work. Evidently there is an immediate opportunity for examining the illustrations derived from steel castings and from hot-worked metal separately.

In this fourth part we will therefore study the former, reserving for the fifth part the study of the latter.

109. In order to place a complete schematic representation of the phenomena we are dealing with upon general foundations underlying every case taken into consideration it would be necessary to take into account the special characteristics shown by the crystalline structure of the steel, within whose elements occur the processes of diffusion and segregation which we have studied from a general point of view in the first two chapters.

tallography" and would obviously oblige us to analyze a great number of phenomena similar to those which have already been incidentally mentioned, such as cold-working, plastic deformation, or twinning, which although of great technical importance among those factors determining the physical properties of steel, are not immediately related to the phenomena of diffusion. As has already been emphasized, diffusion is the only one, therefore, to which it is necessary that we restrict our study in order not to trespass the limits which we have set.

But the facts which must be related as concrete examples illustrating the processes which up to the present time have been studied more or less abstractly could not be clearly comprehended should all considerations concerning their crystalline morphology be disregarded. It is therefore necessary to mention in advance some brief observations concerning such properties, but they will be strictly limited to those necessary for the special and restricted scope in view.

Thus, for our limited purpose, it is unnecessary to take into account the phenomena of "large crystallization" (so-called by Belaiew). Certain factors characterize and accompany the formation of huge crystallites or the large primary crystalline colonies separating from molten steel. They also determine the variations in structure, chemical composition, and physical and mechanical properties in various regions of a large mass of steel, by reason of an accompanying wholesale segregation. Their analysis constitutes one of the most important parts of the general study of crystallization of steel. Yet they produce effects which generally are spread over limits of space much greater than the distances which in practice may be influenced by the processes of diffusion.

It is proper, however, to clarify the limits of that group of phenomena as being those whose effects are not perceptibly modified by ordinary heat treatments. As an example, one might mention those causes which are responsible for the particular distribution of the large crystalline colonies characterizing the various types of steel ingots. It is known that the course of such phenomena is responsible for the forms and the customary positions of the pipe, the presence and distribution of the cracks or "checks" so-called in various types of ingots, the value and distribution of chemical segregation, etc.

**110.** From what has been said in the previous paragraph it follows that for our particular purposes our considerations concerning the crystallography of steel may be restricted merely to a brief examination of the forms assumed in various instances by the first crystallites solidifying during primary crystallization, disregarding the particular way in which these elements arrange themselves to form the large dendrites.

In order better to fix our ideas upon a concrete example, consider a

principal axis of the large dendrites is different in various regions of the metallic mass. Due precisely to that different disposition of the large dendrites various parts must be distinguished in the steel presenting special traits in some property, such as, for instance, the tendency to form forging or rolling checks, the distribution and the nature of the impurities, etc.

Figure 37 shows schematically the arrangement of the principal axis of the large dendrites in one of the innumerable cases which may happen in practice. Without engaging in a detailed examination of this case, which would be out of place here for reasons already mentioned, even a rapid and superficial examination would easily distinguish four regions, distinct one from the other in the respects just mentioned. These regions are the following:

A. The principal axis of the large dendrites in the peripheral region marked with the letter *A* are arranged in concordant directions approximately normal to the external surface of the ingot. This region has a quite variable depth, other conditions being equal, for steels of different composition. In some steels, and especially in ingots of small dimension in which the metal solidifies rather rapidly, region *A* is limited in thickness to a few inches, and sometimes even to a small fraction of an inch. In other steels, and especially in large ingots which solidify slowly, it may reach a thickness of from several inches to one foot or more, even extending sometimes to the axis of the ingot, eliminating entirely the region *B*.

Figure 38 furnishes an example of this case. It reproduces at about one-third natural size the appearance of the plane surface obtained by cutting along its axis a cold ingot of medium hard 2-per cent. nickel steel, subsequently etching the polished surface for about one hour with a 20 per cent. water solution of sulphuric acid heated to about 60°C. The etched surface was then slightly polished with very fine emery cloth, so that its more protruding parts, corresponding to the axis of dendrites left in relief by the etching appear more definitely in the photograph. It clearly shows large dendrites with their longer axis roughly parallel and at right angles to the periphery extending toward the axis of the ingot and

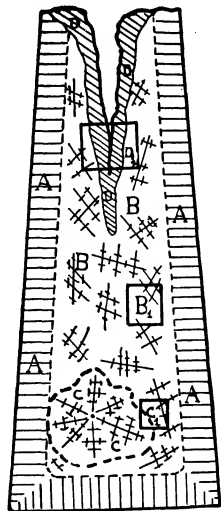


FIG. 37.—Diagram of crystalline orientation in an ingot.

developed tendency for the formation of cracks during hot-work (the so-called "checks"). It is the different development of this region that is responsible for the different tendency of the various steels to such cracks.



FIG. 38.—Split ingot of medium hard 2-per cent. nickel steel.  $\times \frac{1}{4}$ . Pickled in hot dilute  $H_2SO_4$ .

*B.* The region immediately underneath the first one is beyond the columnar structure of region *A*. In it the orientation of the various primary crystallites is promiscuous, occurring in about the same amount along all directions. Its random appearance apparently does not obey any determined law. Figure 39 (prepared and etched in the same manner as Fig. 38) reproduces the appearance of a plane surface in such a region as *B* in a medium hard 1.30-per cent. nickel steel ingot.

It may be pointed out that regions *A* and *B* are not sharply separated

border zone for the larger dendrites to be arranged with their longer axes normal to the periphery.

C. The region *C* in Fig. 37 consists of crystallites surrounding centers of parasitic crystallization. We cannot enter into details of the nature of the reactions developing around these parasitic centers. It suffices

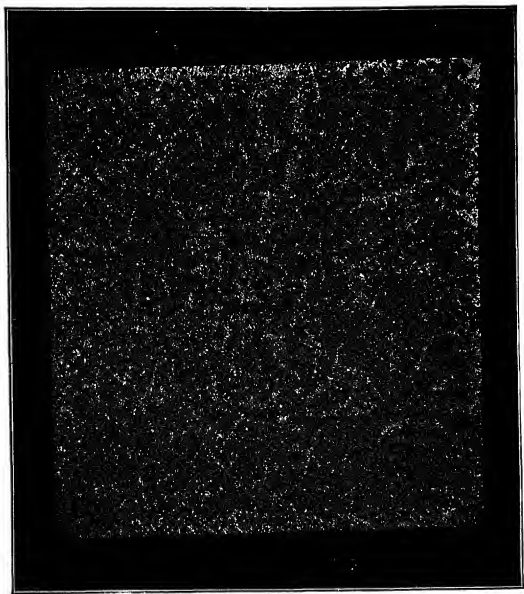


FIG. 39.—Macrostructure of internal region of medium hard 1.3 per cent. nickel steel. Pickled in hot dilute  $H_2SO_4$ .

to remember that parasitic germs form in strongly undercooled liquids. Crystallization occurs around these centers under very different conditions than those holding for the other parts of the ingot. As a result the final crystalline conglomerate has quite different properties from those of the remaining normal metal, both in crystalline structure and in chemical composition, due to wholesale segregation which characterizes this special mode of crystallization.



Wide heterogeneity is clearly illustrated in Fig. 40, which reproduces at about one-fifth of the natural dimensions, a surface<sup>1</sup> obtained by sawing a longitudinal section from an ingot of the same type as the one sketched in Fig. 37. Precisely, Fig. 40 represents an area similar to that marked  $C_1$  in Fig. 37, and includes the surface of separation between region  $C$  and region  $B$ . The different structure of two parts of the steel can be clearly seen, situated respectively at the right and at the left of the separating line crossing the middle part of the figure. In the region situated at the left of the line, the dendritic structure is a great deal better developed than that on the right. The difference between the two structures appears still more evident in the next figure (Fig. 41), reproducing at

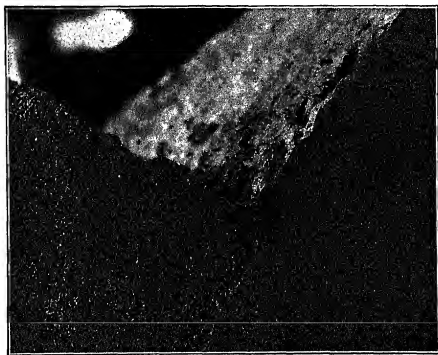


FIG. 40.—Junction between normal and parasitic crystallization. One-fifth size. Pickled in hot dilute  $H_2SO_4$ .

natural size, a more deeply etched part of the same area bordering the surface of separation between the two different crystalline masses.

The steel missing from the upper part of the block photographed in Fig. 40 belonged to yet another crystalline mass of type  $C$ , of Fig. 37, formed around another center of parasitic crystallization. A great amount of impurities accumulated at their contact surfaces as a result of liquation of impure melts of low freezing temperature, and so modified the material immediately at that surface as to allow the metal to break easily along all of it.

constituents of steel, such as carbon, manganese, silicon, etc., but also the greatest part of the impurities which were contained in the molten metal, such as sulphur, phosphorus, solid non-metallic inclusions, etc.

An example of the microscopic appearance which this region may assume is given in Fig. 38.

Similar formations, with the same origin, composition and structure sometimes form in other regions of the ingot, more or less distant from the pipe. A particular example of such a case is given by the region immediately neighboring the surface of separation between the two crystalline masses represented in Figs. 40 and 41.

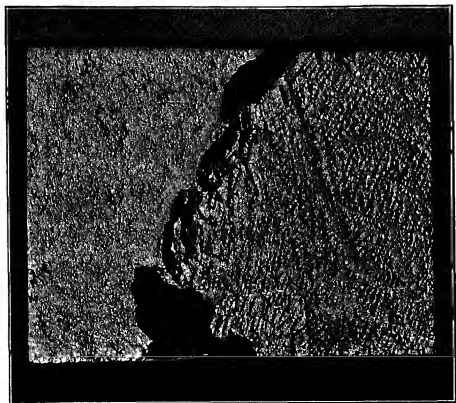


FIG. 41.—Portion of Fig. 40; more deeply etched. Natural size.

**111.** Hereafter we shall disregard the phenomena of "large liquation," with the resulting characteristics sufficiently explained by the examples mentioned above. In subsequent pages other examples will be given to illustrate the variations in normalizing practice induced exclusively by the phenomena of "small liquation;" meaning by this, those segregations which take place between the abutting portions of the primary dendrites.

It may be pointed out immediately that dendritic structure can not be so neatly developed in all steels as in the illustrations presented in the previous paragraph. Nor is it so well marked in the same types of steel under all variations in the conditions under which crystallization and the

successive cooling have taken place. Thus, the development of dendritic structure is, in general, always very imperfect in the very mild carbon-steels. As an instance, the steel ingot shown in section in Fig. 42 was polished and etched under the same conditions as were the surfaces

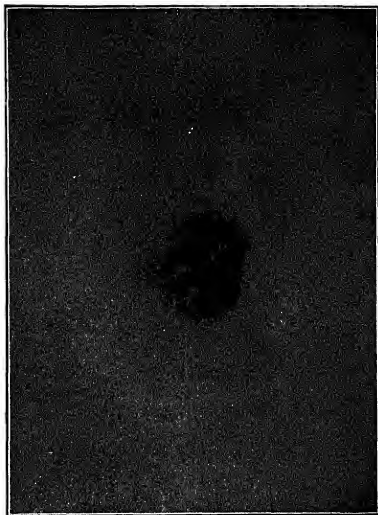


FIG. 42.—Center of 0.8-carbon steel ingot after pickling 1 hour in hot dilute  $H_2SO_4$ .

reproduced in Fig. 38, 39 and 40. Its chemical composition was as follows:

Carbon.....	0.08 per cent.
Manganese.....	0.58 per cent.
Silicon.....	0.07 per cent.

No traces of dendrites appear upon the surface thus treated, and not even a much deeper etching is able to reveal well developed true dendritic structure. This last fact is shown clearly in Fig. 43, which reproduces the appearance of the same surface after the etching with hot dilute sulphuric acid has been repeated a second time, prolonging it for three hours.

112. This experiment establishes, although in this case, and in subsequent

nomena of diffusion take place, it would seem logical that the selection of actual examples chosen to illustrate the effects of normalizing practice should begin with those in which the phenomena studied in Part I would be isolated. In other words, it would be natural to study by concrete examples those phenomena which intervene only in the liquation processes accompanying primary crystallization of steel and the processes of diffusion in the austenite, without being modified by the phenomena concomitant with the transformation of gamma to alpha iron.



FIG. 43.—Same surface as Fig. 42 after etching for 3 hours.

However, such a proceeding would immediately lead one outside the field which strictly limits the present study; namely, the field of the soft and medium steels normally used for machinery parts.

In fact, in the present state of the art, mechanical construction never employs steels in which the  $\gamma \rightleftharpoons \alpha$  transformation has not taken place in a larger or smaller measure. Even those extremely rare cases in which some austenitic steels are used for certain machine parts do not constitute an exception to this rule, for those parts are not mechanical parts properly

logical characteristics manifested by the phenomena of the second group as they are related to those of the first. In other words, it is necessary to examine briefly in what forms the crystalline texture resulting from the  $\gamma \rightleftharpoons \alpha$  transformation superposes upon the more or less imperfect system of dendrites resulting from solidification.

**113.** This second aspect of the crystallography of steel, if studied in a thorough way and in all its details, would constitute one of the most important chapters of our knowledge. The relations between the phenomena due simply to the processes of diffusion and the phenomena originating from processes of purely crystallographic character are here no less rigorous than for primary crystallization. Therefore, one must evidently confine his discussion of this group to that which is strictly necessary for the right understanding only of those facts which we will have to examine further on.

Part II contained a general statement of the *normal* relations existing between the course of the primary separation of  $\gamma$  mixed crystals from molten steel and the secondary separation of the  $\beta$  and  $\alpha$  mixed crystals from the austenite. Part III also mentioned in a general way, two of the principal groups of phenomena which interfere frequently in practice, and disturb those normal relations (see Secs. 93 to 107).

Therefore, let us now see some examples of the morphological aspects which those relations may assume.

It is clear that the simplest case which may be presented is that where-in the diffusion of the various elements throughout the austenite has taken place so completely as to cause perfect homogeneity of concentration.

As already often noted, such an extreme and perfect case seldom or never occurs in industry. Nevertheless whenever it should happen, it is evident that the successive secondary crystallization of the  $\beta$  and  $\alpha$  crystals would take place freely, independent from the previous course of the primary crystallization and according to the laws studied in Part II, except, naturally, as a consequence of other eventual perturbative conditions such as these noted at the end of Part III.

However, it is possible to mention some examples in which the uniformity of concentration reached in the  $\gamma$  mixed crystals approaches very near the extreme limit. In such cases the secondary crystallization will furnish very close approximation of the phenomena as they should take place in that extreme theoretical case.

**114.** Figure 44 reproduces at 30 dia., after polishing and etching with a 5 per cent. solution of picric acid in alcohol, a section of a steel plate from a cementation muffle. This plate was maintained during more than

At the end of this period the furnace fires were drawn and the plate cooled, together with the entire mass of the furnace, so that the temperature dropped very slowly and uniformly, passing from about 1000°C. down to 150°C. in a period of time longer than one hundred hours.

It is also necessary to observe that, while one of the faces of the plate was constantly in contact with the carburizing substances charcoal and carbon monoxide, the other face was bathed by oxidizing gases. Naturally, this circumstance provoked a continuous migration of carbon during the entire life of the muffle, due to the diffusion of dissolved carbon compounds through the entire thickness of the plate, from the inside to the outside. There is no doubt that this continual transfer must have con-



FIG. 44.—Alpha iron in cementation muffle.  $\times 30$ . Etched in alcoholic picric acid.

tributed largely to the attainment of a high degree of homogeneity in the carbon concentration in each crystalline element of the  $\gamma$  solid solution.

It is clear that the circumstances just described may be considered as among the most favorable which can be realized in practice to obtain a thorough uniformity of the concentration of carbon in all parts of the individual crystals of austenite. In such a case, therefore, the secondary crystallization should proceed with the least interference from pre-existing conditions inherited from primary crystallization. It should place in evidence in the best possible way, the structural characteristics resulting from the mere separation of  $\alpha$  ferrite during allotropic transformation.

In the case taken as an example, this last result is obtained in a manner

quite close to that proceeding at the theoretical limit of perfect homogeneity of austenite, inasmuch as the original average concentrations of the elements other than carbon contained in solid solution in the  $\gamma$  iron, were very low. Consequently their eventual residual heterogeneities in the steel after its prolonged heating must necessarily be extremely small, an important point, since these heterogeneities have been seen to exercise appreciable effects upon the separation of ferrite. In fact, the composition of the steel constituting the plate observed as far as the other four fundamental elements beside iron and carbon are concerned was as follows:

Manganese.....	0.40 per cent.
Silicon.....	0.01 per cent.
Sulphur.....	0.01 per cent.
Phosphorus.....	0.05 per cent.



FIG. 45.—Rounded alpha iron and lamellar beta iron from homogeneous austenite.

We may therefore, consider that the characteristic structure of the crystals of  $\alpha$  iron reproduced in Fig. 44 is at least closely approximate to that peculiar to ferrite separated from  $\gamma$  mixed crystals of absolutely uniform concentration.

It is known that such so-called Widmanstätten structure is built up of flat plates of ferrite, oriented according to definite directions. They satisfy the crystallographic criteria showing them to be arranged in a family of planes parallel to the surfaces of the octahedron, one of the simpler crystalline forms of the isometric system. Their mutual intersections vary according to the direction which the plane of the polished

Another observation may be made even upon Fig. 44, although the fact is shown still clearer in Fig. 45. The latter is a microphotograph of the structure at another point of the same section of plate, selected where the carbon is lower. In fact, the specific carbon content varies from a very high value at the internal surface of the muffle, to almost zero at its external surface, a state of affairs due to the history of this plate, which has been described.



FIG. 46.—Fracture of a steel having highly developed Widmanstätten structure. Natural size.



appears in the form of the well known roundish kernels characteristic of soft steels. On the other hand in regions where the concentration of carbon exceeds the value of about 0.4 per cent., ferrite appears in the form of needles oriented along octahedral cleavage (Widmanstättian structure). For a series of reasons to which I cannot here refer,<sup>1</sup> the hypothesis appears justified that the granular appearance of the crystals in low-carbon areas is that characteristic of  $\beta$  ferrite. Carbon steels containing less than 0.4 per cent. carbon separates  $\beta$  iron directly from  $\gamma$  mixed crystals (see Fig. 10, curve *DF*) and this is transformed afterwards at lower temperatures (along curve *BC* of Fig. 10) into  $\alpha$  iron or ferrite without a substantial change in its microscopic appearance. The lamellar form of the



FIG. 47.—Widmanstättian needles in grains definitely outlined by reticular structure.  
× 50.

crystals in higher carbon steels, on the other hand, would possess the characteristics of the crystals of ferrite, or  $\alpha$  iron, which separates *directly* from the  $\gamma$  solid solution (see curve *FH* of Fig. 10).

115. From the point of view of mechanical properties Widmanstättian structure in hypoeutectic steels is known to be associated with high brittleness. Their fracture appears brilliant and coarsely crystalline. Fig. 46 reproduces at about natural size the appearance of the fracture of a soft steel having this structure highly developed.

Such an abnormal development very seldom occurs and only in case of very prolonged and strong overheating, so as to approach the region of true burning (see Sec. 25, 26 and 83). Less severe instances of over-

heating happen quite frequently in practice, especially in the heating operations preparatory to rolling or forging. Even this is always such as seriously to impair the mechanical properties of the steel. In such cases the structure of the resulting metal shows a combination of the pure needle-like Widmanstätten structure with the reticular network structure which shall be examined shortly. An example of such a mixed structure is that reproduced in Fig. 47, enlarged 50 dia.

However, in such intermediate cases we are obviously confronted with phenomena of secondary crystallization occurring in a  $\gamma$  solid solution which is no longer even approximately homogeneous. These cases, therefore, do not belong in the category with which we are now concerned. We shall have occasion to refer to them further on.

As far as concerns the extreme cases of homogeneous austenite mentioned in the previous section, I may note that it is nearly always possible to modify the pure Widmanstätten structure, into a mixed structural type similar to that of Fig. 47 and of improved mechanical properties, by disturbing or interrupting the crystallization of  $\alpha$  iron, either by a rapid cooling through the critical interval, or by mechanical work through the upper part of the same range. The details of this process of disturbed crystallization does not concern us here because the extreme cases of uniform homogeneity such as the one related above have but a petty industrial importance.

## CHAPTER XV

### ANALYSIS OF THE PRELIMINARY HEAT TREATMENT OF A SPECIAL STEEL

116. Passing by the theoretically important limiting case of completely homogeneous austenite, let us examine the structural evidence appearing in an actual steel during the separation of  $\gamma$  mixed crystals from the melt and its later disintegration into  $\beta$  and  $\alpha$  crystals. By exclusion this discussion will relate to the only industrially important and frequent case in which the  $\gamma$  mixed crystals preserve to a larger or smaller degree the heterogeneities of concentration in their various constituents produced during primary solidification in the manner indicated in Part I.

To this effect let us first analyze one of the cases in which the effects of those relations are strongly apparent. Such conditions occur prominently in a medium hard 2-per cent. nickel steel, similar to the one whose microstructure is reproduced in Fig. 38.

Let us start by examining this steel in the raw state, as cast; in other words in the condition of maximum heterogeneity in austenite.

The composition of this steel was the following:

Carbon.....	0.42 per cent.
Manganese.....	0.60 per cent.
Silicon.....	0.22 per cent.
Sulphur.....	0.008 per cent.
Phosphorus.....	0.02 per cent.
Nickel.....	2.02 per cent.

The dendritic structure revealed by warm dilute sulphuric acid attack was identical to the one reproduced in Fig. 38.

Let such a section of the same metallic surface be accurately repolished and lightly etched with one of the usual metallographic reagents for the examination of hypo-eutectoid steels (for instance, 5 per cent. picric acid solution in alcohol). If it is then examined, that ensemble of structural characteristics which H. M. Howe has called ingotism will distinctly appear in the areas formerly exhibiting normal dendritic structure, corresponding to regions A and B of Fig. 37.

Figure 48 reproduces at 3 dia. this exact structure as it appears near the surface of a 1-ton ingot of the steel in question, after polishing and etching with a 5 per cent. solution of nitric acid in amyl alcohol.

The micrograph reveals clearly the characteristic veins or shells of

is easy to understand how such a distribution of the constituents corresponds to the course of secondary crystallization, where the  $\alpha$  solid solution precipitates from the non-uniform  $\gamma$  mixed crystals, as has been carefully analyzed in Part II, Sec. 63 to 74. Dr. Howe's new word "ingotism" describing this structure is very suggestive of the characteristic structure of ingots, and it is a word which has therefore been generally adopted. The veins of ferrite, constituting a real network, are deposited by the successive nuclei of  $\alpha$  iron, primarily formed at or very near to the primary crystallization centers—that is to say, region  $\beta$  of Sec. 66, where the concentration of carbon and other elements was the minimum—and result from the successive deposit of lamellæ of  $\alpha$  iron upon these germs. The islands of ferrite contained in the interior of the principal meshes (that is, in the kernels) are a growth around new crystallization germs of  $\alpha$  iron, appearing successively in the more advanced stages of the  $\gamma \rightarrow \alpha$  transformation range. The mechanism was discussed under "region C" of Sec. 66. Moreover, the structure which is now being examined constitutes a typical example of the forms assumed by the  $\alpha$  crystals in the general case of their separation from the heterogeneous  $\gamma$  solution. There is an evident tendency for crystalline elements which successively build themselves upon pre-existing germs of the same nature to arrange themselves in ranks along prevalent directions, so as to result

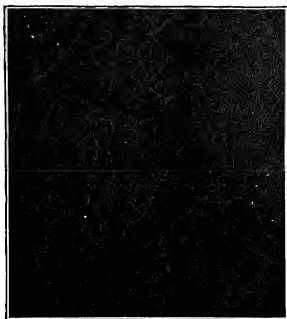


FIG. 48.—Thin ferrite network in nickel-steel casting.  $\times 3$ . Etched by 5 per cent.  $\text{HNO}_3$  in amyl alcohol.

in long tiny branches rather than massive accumulations. This fact is entirely similar in its causes and effects to that which is responsible for the formation of dendrites during primary crystallization of the  $\gamma$  solid solution. It also better defines from a morphological point of view the phenomena studied in Part II, especially Sec. 66 and 67.

In order to exclude the hypothesis that the characteristic disposition of reticular structure which has just been examined might be due to the presence of non-metallic inclusions capable of localizing and orienting the growth of the  $\alpha$  crystals in the manner which has been indicated at the end of Part III,<sup>1</sup> I may immediately state that the islands or veins of ferrite do not contain non-metallic inclusions except in some rare instances

<sup>1</sup> See especially Sec. 98.

in the material which we are now concerned with and illustrated in Fig. 48. This can be seen clearly in Fig. 49, which reproduces at higher magnification (80 dia.) a spot on the same metallic surface.

117. Two different methods of approach may be adopted in order to examine the relative disposition of the two crystalline networks,  $\gamma$  and  $\alpha$ , resulting respectively from the primary and the secondary crystallization. I think it opportune to mention an example of each of them, referring always to the same steel, whose characteristics have been indicated in the previous section.

The first method consists in etching the polished surface with hot dilute sulphuric acid, until the dendritic system is well developed. The pitted surface is then slightly polished to prepare the surface in relief



FIG. 49.—Ferrite vein free from sonims.  $\times 80$ .

for a subsequent etching with a 5 per cent. solution of nitric acid in amyl alcohol, prolonged only long enough to place the ferrite in evidence.

Figure 50 reproduces at 10 dia. the structure of a piece of the nickel steel doubly polished and etched in the manner just described. As is shown, there exists only a general concordance between the "habit" and the orientation of the two structural systems constituted respectively by the dendrites—crystals which originally were formed by  $\gamma$  iron—and by the elements of ferrite—crystals of the  $\beta$  and  $\alpha$  series. Certainly, the veins of ferrite in the network do not correspond uniformly to elements of the system of the dendrites as determined by deep etching.

The conclusion is evident that the orienting forces responsible for the formation of the ferrite network and dendrites in the  $\beta$  ( $\alpha$  and  $\beta$ )

and sometimes overwhelms the forces due to difference in concentration of the various elements in the heterogeneous austenite. For evident

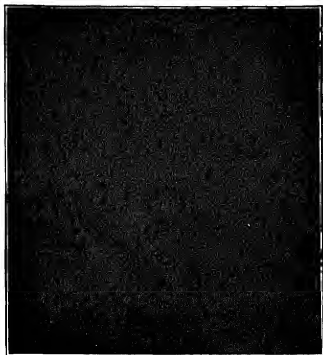


FIG. 50.—Nickel steel after double etching.  $\times 10$ .



FIG. 51.—Nickel steel etched with  $\text{HNO}_3$  in amyl alcohol.

reasons, it is not possible to tarry and analyze these phenomena as they appear in Fig. 50 and in all the analogous cases.

The same facts are revealed even in a clearer way when the second of the two methods of observation is applied. This consists in photographing a given spot on a polished surface after etching it with the amylic solution of nitric acid so as to reveal the ferrite network. Then exactly the same area is photographed a second time after being etched with the hot sulphuric acid solution, which develops the primary dendrites. Care is taken to make the two photographs at the same enlargement. Then place the two negatives exactly one upon the other with precautions which are not necessary to describe here, and print through the two so as to get a single positive. This print reveals distinctly and



FIG. 52.—Fig. 51 after pickling in hot  $H_2SO_4$ ; photographed at same magnification.



FIG. 53.—Superposed negatives, Fig. 51 and 52.

completely the reciprocal positions of each structural element of the two systems.

The three figures, 51, 52 and 53, representing a piece of the same 2-per cent. nickel steel, were made in accordance with the methods just described. Figure 51 reproduces the appearance of the metallic surface simply etched with an amylic solution of nitric acid. Figure 52 is the same surface etched with hot sulphuric acid. The third, Fig. 53, is obtained by placing one negative upon the other.

In this case we may also repeat the observations already made previously but with a greater degree of certainty because it is no longer

have happened in the former case. In particular, we see that, if in some regions of the surface the continuous veins of ferrite seem to accompany the pitted zones separating the principal branches of the dendrites, in a far greater number of cases this relationship does not hold. We find, therefore, a confirmation of the conclusion which we had already reached concerning the greater influence of the orienting forces responsible for the ferrite network compared with the influence of the concentration-differences in the various regions of the  $\gamma$  mixed crystals.

**118.** Now that the morphological relations existing between the two crystalline systems are fixed in this way, at least in outline, it will be interesting to add some observations regarding the behavior of their respective elements when under sufficient stress to permanently deform or fracture the steel. Thus will be seen better the role played by the elements of each of the two systems whose properties are superposed to form the sum total of the peculiar mechanical properties characterizing ingotism in raw steels, and especially medium hard ones. Of these properties we shall see some examples further on.

From an unworked ingot of the same 2-per. cent. nickel steel, was taken a flat test-piece of the well known type proposed by Frémont. One of the plane faces of this test-piece was polished, but only partially etched in a longitudinal strip by spreading over the metallic surface a large drop of the reagent, in this case a 5 per cent. solution of nitric acid in amyl alcohol. Figure 54 shows the appearance of this face after fracture in tension.

The following facts are clearly seen when the deformed region of the surface is examined, and above all when the prolongation of the veins of ferrite comprising the large network is followed attentively in the unetched part:

1. The surfaces along which fracture has taken place totally, or even only initially, correspond entirely to the large network of ferrite.

2. The ferrite shells do not seem to have any relation at all to the slip bands or lines of Luders. The latter appear in the bright unetched part of Fig. 54. It is easy to see among other things that the great majority of the slip bands have great deal smaller dimensions than the meshwork of ferrite and are totally contained within the kernels but remain independent of the shells.

It is therefore apparent that fracture under slowly applied stress takes place along surfaces of less resistance closely connected to the ferrite meshwork before that same fracture may occur at the joints of the structural system related to the slip bands. This remark applies to a medium hard steel, as cast, of the same type as the one examined, in which ingotism is highly developed.

Now many considerations which cannot be recounted here suggest



dendrites. We shall see presently in the description of similar cases, that preliminary heat treatments very easily obliterate the large network of ferrite, but they must be pushed to an extreme and executed with particular care in order to attenuate the dangerous effects of the dendritic structure. Consequently one may predict that the effects of normalizing upon the mechanical properties of steel are felt according to two distinct *degrees*. The first one corresponds to the obliteration of the large ferrite

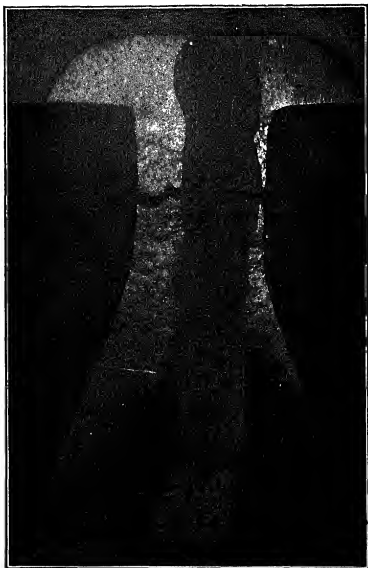


FIG. 54.—Broken Frémont test piece.

network, and can be reached easily, even by mild treatments; while the second one corresponds to a modification of the characteristics of the dendritic system, and may be obtained only by very energetic homogeneity treatments.

Of these facts we shall also see some examples later.

In order to do this I will first summarize briefly the results of some observations made upon the same 2-per cent. nickel steel discussed in the previous paragraphs.

A square ingot of this steel  $15\frac{3}{4}$  in. on a side was cast, and several prismatic test-pieces  $1\frac{3}{8} \times 1\frac{3}{8} \times 7\frac{7}{8}$  in. in size were cut. Twelve of these test-pieces were submitted to the heat treatments indicated in Table I. The annealing temperature was reached in about one hour and a half from the cold. Slow cooling in the furnace was done so that the temperature of the test-pieces lowered from the high heat to 200°C. in about six hours. Air cooling was done by suspending the test-piece so as to allow the air to circulate freely around it.

TABLE I.—SCHEDULE OF HEAT TREATMENTS

No.	Identification marks <sup>1</sup>	Heat treatment		Brinell hardness (under 3000 kg.)
		Heating	Cooling	
1	$A_3$	1 hour at 700°C.	Slowly, in the furnace	179
2	$A_{3-1}$	1 hour at 700°C.	In the air	179
3	$B_3$	2 hours at 700°C.	Slowly, in the furnace	179
4	$B_{3-1}$	2 hours at 700°C.	In the air	183
5	$D_3$	1 hour at 800°C.	Slowly, in the furnace	179
6	$A_2$	1 hour at 800°C.	In the air	179
7	$B_2$	2 hours at 800°C.	Slowly, in the furnace	179
8	$B_{2-1}$	2 hours at 800°C.	In the air	183
9	$D_4$	1 hour at 900°C.	Slowly, in the furnace	183
10	$D_{3-1}$	1 hour at 900°C.	In the air	183
11	$D_{4-1}$	2 hours at 900°C.	Slowly, in the furnace	179
12	$C_{1-1}$	2 hours at 900°C.	In the air	192

<sup>1</sup> The conventional marks contained in the second column have been included because they indicate the corresponding test-pieces in some of the photographs reproduced further on.

From the microscopic examination of each reheated test-piece after double etching as described in Sec. 117, the following facts were ascertained:

1. None of the twelve heat treatments has obliterated nor even appreciably modified the dendritic structure of the steel, as revealed by sulphuric acid attack. For example, Fig. 55 reproduces at an enlargement of 3 dia. the surface of test-piece No. 12, polished and etched with hot dilute sulphuric acid.

As is evident, the dendritic structure still appears clearly in this test-piece even after being subjected to the longest heating (2 hours)

2. In the first four test-pieces heated to 700°C. the heat treatment has not modified the ferrite network to an important degree. It continues to appear as large meshes, whose elements are entirely independent of the dendrites revealed by the sulphuric etching. For example, Fig. 56 reproduces at an enlargement of 65 dia. the appearance of test-piece No. 3 after being subjected to double etching under the conditions indicated at the beginning of Sec. 117.

This result is in accordance with the position of the first transformation point shown on heating this steel. In fact,  $A_{c1}$  lies above 700°C. as is seen from Fig. 57, the differential heating curve traced with the Le Chatelier-Saladin apparatus.

3. Starting with the fifth sample, the heat treatment impresses itself strongly upon the structure of the ferrite. Eutectoid cementite passes into solid solution in  $\gamma$  iron when the initial temperature of allotropic transformation on heating has been exceeded. Upon cooling, crystals of ferrite separate anew from this solution where the (diffusion phenomena studied in Part II have taken place continually) forming the new network whose characteristics evidently depend upon the length and temperature of heating and upon the

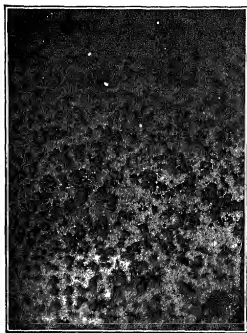


FIG. 55.—Test-piece 12 after etching with hot dilute  $H_2SO_4$ .

velocity of subsequent cooling.

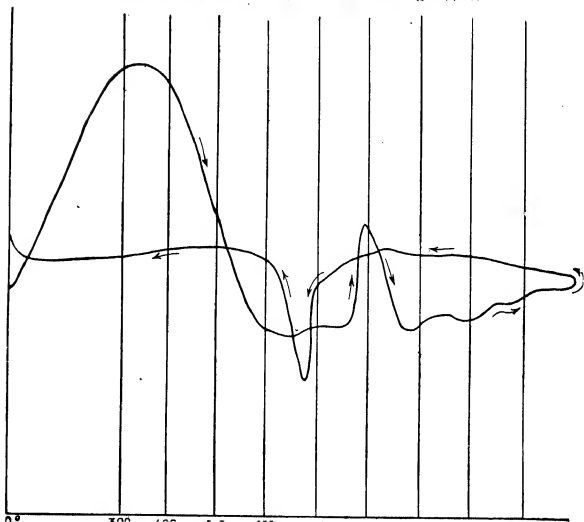
A complete comparative examination of the microstructure of test-pieces No. 5 to 12 inclusive must be omitted for the sake of brevity.<sup>1</sup> It shows that the ferrite meshwork formed after the heating is the wider the higher has been the annealing temperature, which indeed is in perfect accordance with the argument of Part II. Also in accordance with the theoretical considerations, it is a fact that the ferrite meshwork in the reheated steel is the closer, the more rapid has been the cooling, heating conditions being equal.

For instance, this last fact is clearly shown upon comparing the structure of test-piece No. 11, reproduced in Fig. 58 at 65 dia., with

<sup>1</sup> This has been given by the author in *Chemical & Metallurgical Engineering*, Vol. 22, p. 921, May 19, 1920.



FIG. 56.—Test-piece No. 3, after double etching.  $\times 65$ .



that of test-piece No. 12, reproduced with the same enlargement in Fig. 59.



FIG. 58.—Nickel steel after 2 hr. at 900°C. and cooling in the Furnace. Double etched.  $\times 65$ .

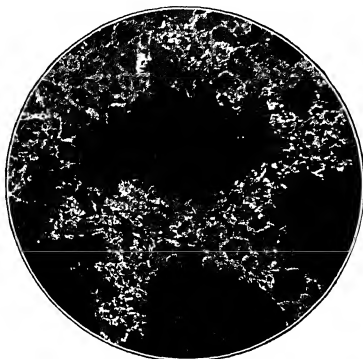


FIG. 59.—Nickel steel, 2 hr. at 900°C. and air-cooled. Double etched.  $\times 65$ .

It is interesting to record the fact—which shall presently be seen to have considerable practical importance in determining the mechanical

heat treatments even less energetic, such as a simple heating followed by a normal cooling, are capable of transforming a very coarse network of ferrite in a very fine one.

**120.** Before proceeding further in the examination of the effects of normalizing practice upon the steel we are studying, I might call attention immediately to an observed fact brought out in the preceding section, and which is easily explained on the basis of the considerations of Part II.

Comparing the microstructures of our nickel steel as cast (reproduced for instance, in Fig. 48, 49, 51 and 56) with that of the same steel after reheating at temperatures above  $780^{\circ}\text{C}$ . the point  $A_{c1}$  (reproduced, for instance, in Fig. 58 and 59), one notices immediately that the pro-



Fig. 60.—Structure of a 0.33-per cent. carbon steel near surface of the ingot. Etched with picric acid in alcohol.  $\times 60$ .

portional area occupied by the ferrite and the pearlite respectively is quite different. The proportion of ferrite is much greater in the reheated steel than in the raw steel. In the sections reproduced in Fig. 48 and 49 this proportion is a great deal less than that which would theoretically result on the basis of the carbon in the steel as determined by chemical analysis and from the application of known criteria from the iron:carbon equilibrium diagram, where the pure eutectoid is obtained at carbon equal to 0.9 per cent.

This fact is not unique for a special steel of this type. Instead, it is very general in character and appears in greater or lesser measure in almost all the medium steels, whether straight carbon steels or more

Just to give an example selected from among the more usual carbon steels, Fig. 60 represents the structure of a medium steel, having the following composition as determined from a test-piece taken from the ladle after teeming was half completed:

Carbon.....	0.33 per cent.
Manganese.....	0.57 per cent.
Silicon.....	0.23 per cent.
Sulphur.....	0.02 per cent.
Phosphorus.....	0.03 per cent.

The structure reproduced in Fig. 60 (enlarged 60 dia. after etching with an alcoholic solution of picric acid) is that exhibited at a depth of about 5 mm. below the external surface of a 2-ton ingot. The carbon

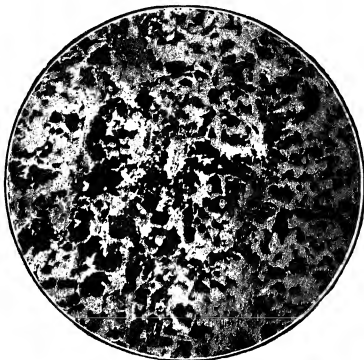


FIG. 61.—Same steel as Fig. 60, after annealing 3 hr. at 850°C.; and slow cooling.

analysis, made in that precise region of the metal showed carbon to be 0.31 per cent. The presence of such local variation of carbon in that region of the ingot which has cooled first is in perfect accord with the considerations developed in Part I.

The same sample of steel heated for 3 hours at 850°C. and left to cool slowly, and then re-examined in the same region gave the structure reproduced in Fig. 61.

Note that the second photograph was made after having cut away a layer of metal more than 4 mm. in thickness, so as to exclude the possible effects of any eventual decarburization due to chemical action of furnace

when the fact is known that an analysis made after the microscopic examination and at the precise point observed did not reveal any loss in carbon.

Contrary to what was observed in the steel as cast, it seems clear from Fig. 61 that in the reheated steel the proportional quantity of ferrite to pearlite is exactly that deduced from the iron:carbon diagram at complete equilibrium. A steel containing 0.31 per cent. carbon should show a little less than two parts of ferrite to one part of pearlite.

The same set of facts was found in the deeper regions of the ingot, in which the carbon concentration is higher, for reasons developed in Part I.



Fig. 62.—Same steel as Fig. 60, but 50 mm. below surface.  $\times 60$ . Etched with picric acid in alcohol.

Thus, Fig. 62, under the same enlargement and etching of the two previous, gives the structure of the same steel ingot at a depth of about 50 mm. below the surface. A carbon analysis in this precise region gave carbon as 0.39 per cent. As shown by the half-tone the proportion of ferrite is greater, although the concentration of carbon is higher in this region where the steel has cooled slowly, than in that further toward the surface.

In the same steel, an annealing for 3 hours at  $850^{\circ}\text{C}.$ , followed by a slow cooling, gives rise to a modification of the relative proportion of metals entirely similar to that observed in the previous case. Figure 63 shows the structure of the same test-piece as Fig. 62, after having undergone an annealing as described and with the precautions already indicated for the test-piece reproduced in Fig. 61.



an example it suffices to remember how frequently in mill practice nowadays one estimates the approximate variation in carbon content of steel from point to point by a simple microscopical examination. It is easy to imagine the errors induced by the phenomena we have described in the very frequent instances where the examination is made upon a cast medium steel.

An explanation may be found among the considerations developed in Part II. A review made from the special view point of the phenomena which we are now concerning ourselves with, would clearly evidence the great complexity of their underlying causes, but on the other hand, would

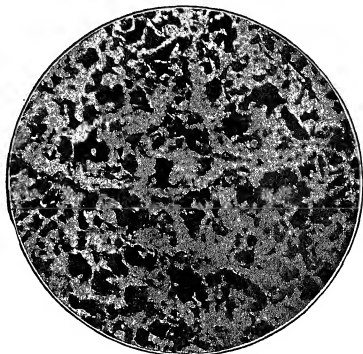


Fig. 63.—Same as Fig. 62 after 3 hr. at 850°C. and slow cooling.

demonstrate that the ultimate origin always goes back to the variable concentrations of carbon and the other elements in solid solution in  $\gamma$ ,  $\beta$ , and  $\alpha$  iron, and to the variations which these heterogeneities undergo owing to the various course of diffusion taking place in the steel during its solidification, its further cooling, and during the various heat treatments.

It is not possible to recapitulate the argument because it would entail too long a diversion upon matter which is only indirectly related with the main purpose of our study. Therefore, only some references to Part II will be mentioned wherein a direct examination of the phenomenon of suppressed ferrite has been given.

Thus, one might read the remarks made in Sec. 47 regarding the

effects of changing the cooling velocity upon the proportion of ferrite appearing in a hypoeutectoid steel together with those of paragraph *b* of Sec. 51 concerning the dependence of the carbon concentration in  $\beta$  and  $\alpha$  mixed crystals of a given kind upon the temperatures at which this series forms from heterogeneous austenite. It is also necessary to take into account the observations made in paragraph *c* of the same section.

The remarks made at the end of Sec. 54 are equally important for a complete explanation of the phenomenon now under discussion. This reference concerns the dislocations in the curve of maximum carbon-concentrations in austenite suffered by a variation in the frequency of crystallization germs of  $\alpha$  and  $\beta$  phases. The end of paragraph *b* and paragraph *c* of Sec. 61 also bears on the matter, in what is said there of the similar effects of surfusion occurring in certain regions of the  $\gamma$  solid solution during its transformation.

Some of the causes of ingotism are better determined and explained by the remarks made in Sec. 62, 65, and 66, particularly in the last part of each of them. This especially so far as they apply more directly to the differences between the proportion of ferrite contained in a steel which is quite non-uniform at the instant of the  $\gamma \rightarrow \alpha$  transformation and the amount of ferrite contained in another steel in which these heterogeneities have been more or less smoothed out by means of a homogeneity heat treatment.

In order to sum up in a few words the generic characters of all causes of ingotism, isolating the more important group of phenomena from the great number of the concomitant secondary facts analyzed in the preceding pages, this is what could be said:

Sections 63 to 68 showed that as the heterogeneity of the concentration of carbon and the other elements in the  $\gamma$  mixed crystals is greater—that is to say, as the band of variant concentrations is wider—the lower is the concentration of carbon in the  $\beta$  and  $\alpha$  mixed crystals which separate at the beginning of the  $\gamma \rightarrow \alpha$  transformation. Consequently, the lower will be the proportion of carbon which is subtracted from that residue of austenite which afterwards forms the eutectoid pearlite of practically fixed composition. It follows from this, that each decrease of the degree of heterogeneity of the concentrations in the  $\gamma$  mixed crystals gives rise to a decrease in the carbon available for the formation of the eutectoid and, therefore, to a decrease of the proportion of the eutectoid itself.

It is true that an increase in the width of the band of heterogeneity in the austenitic field may cause an increase in the proportion of  $\alpha$  mixed crystals richer in carbon which form at the end of the transformation, but it is clear that there is no necessary proportional relation between the

previous section demonstrate simply that the influence of the first group exceeds that of the second, at least in the cases examined.

Beside all this, it is necessary to take into account the demonstrated fact that the variations in concentrations of the various elements (principally carbon) throughout the austenite may cause the formation of many extremely small crystallites of ferrite, as is clearly seen in the various microphotographs reproduced in Part II and in the last few sections. These crystalline elements remain scattered throughout the eutectoid kernels of the raw steel in many cases, thus escaping recognition and proper appraisal in a microscopic investigation made under ordinary conditions. Yet they are easily seen when they have *agglomerated* in less numerous but larger masses after a reheating. It is clear that such a fact must cause an apparent decrease of the proportion of ferrite in raw steel and an increase (also apparent), of the proportion of pearlite, a change in ratio which accrues to the actual increase which occurs for the reasons just analyzed.

122. Let us now see what variations are produced by normalizing treatments upon the physical properties and the accompanying structural changes in the nickel steel which has been under examination. This is

TABLE II.—MULTIPLE HEAT TREATMENTS

No.	Mark	Operation No.	Heat treatment			Brinell hardness	Corresponding macro-structure	Corresponding micro-structure
			Mode of heating	Quenching in water from	Re-heating			
13	C <sub>1</sub>	..	11 hours at 1090°C. followed by cooling during 2 hours to 800°C.	800°C.	2 hours at 580°C.	217	Fig. 64	Fig. 67
14	A <sub>2-1</sub>	1	9 hours at 1000°C. followed by cooling during 1 hour 45 minutes to 800°C.	800°C.				
		2	9 hours at 1050°C. followed by cooling during 2 hours to 800°C.	800°C.	2 hours at 580°C.	179	Fig. 65	Fig. 68
15	C <sub>2</sub>	1	10 hours at 1190°C. followed by cooling during 2 hours 25 minutes to 800°C.	800°C.				
		2	11 hours at 1160°C. followed by cooling during 2	800°C.				

tions we shall observe the results obtained by subjecting the same steel to extremely energetic homogeneity heat treatments, made under conditions,

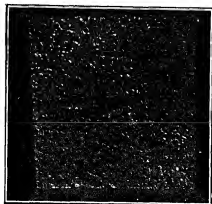


FIG. 64.—Test-piece 13 after pickling in hot 20-per cent.  $H_2SO_4$ .  $\times 2$ .

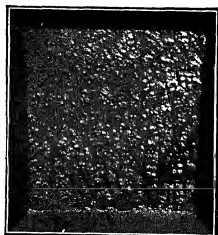


FIG. 65.—Test-piece 14 after pickling in hot 20-per cent.  $H_2SO_4$ .  $\times 2$ .

which, in practice, are considered the most efficacious to strongly transform the original crystalline edifice of raw steel. These results will be used as a fundamental term of comparison, and regarded as the extreme end of a series of most efficacious homogeneity treatments.

In Table II are indicated such treatments impressed upon three test-pieces of the same dimensions and cut from the same 2-per cent. nickel steel ingot. The three test-pieces are numbered 13, 14 and 15 following the previous samples in Sec. 119. For the same reason the distinctive mark showing on each test-piece in later illustrations has been added in the second column.

To avoid misunderstandings it may be pointed out that the two treatments indicated for test-pieces No. 14 were applied successively to the same sample, which was observed after having been subjected to the double treatment. The same remark applies to the three treatments of No. 15.

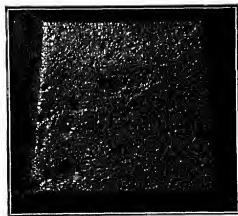


FIG. 66.—Test-piece 15 after pickling in hot 20-per cent.  $H_2SO_4$ .  $\times 2$ .

**123.** Figures 64, 65 and 66 show at 2 dia. the appearance of the three respective test-pieces No. 13, 14 and 15, polished in the usual way, and etched with the hot aqueous solution of 20 per cent. sulphuric acid. Even after the drastic heat treatments to which the first two test-pieces

have been subjected, the system of the dendrites still appears upon etched surface with a clearness practically equal to that with which

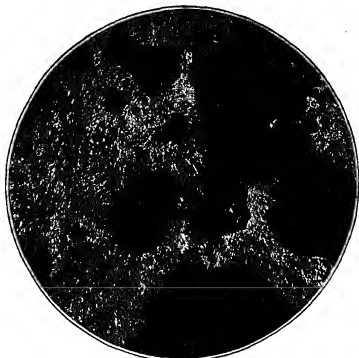


FIG. 67.—Test-piece 13 after double etching.  $\times 65$ .



FIG. 68.—Test-piece 14 after double etching.  $\times 65$ .

double etching is reproduced at an enlargement of 65 dia. in Fig. 67, 68 and 69.

As could be foreseen, there exists no longer a true network of ferrite discernible under medium magnifications but only the usual conglomerate due to reheating at a temperature below  $A_{c1}$ , *i.e.*, a segregation of the solid solution now rendered nearly homogeneous by the previous heat treatments, at least as far as the concentration of carbon is concerned.

Here is a more definite confirmation of the fact that heat treatments designed to foster the phenomena of diffusion and to check or, at least to decrease the processes of segregation taking place during cooling through

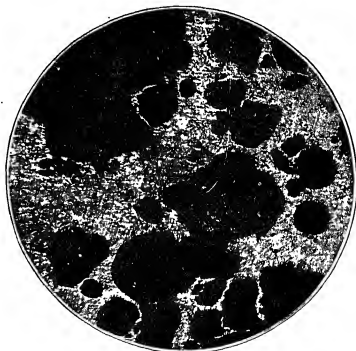


FIG. 69.—Test-piece 15 after double etching.  $\times 65$ .

the transformation interval, make their effects felt with enormously greater rapidity and intensity upon the network of ferrite (or upon the crystalline system which determines its form) than upon the dendritic system.

Experimental data, which would be out of place to dwell upon here, seem to indicate that very remarkable modifications of *form* can be obtained in the structure of the dendritic system, even without excessively drastic heat treatments. However, for a steel of the type under study, it certainly is necessary to subject the metal to extremely energetic heatings and quenchings in order to obtain a perceptible smoothing out of the local heterogeneities, by which the dendritic system is revealed by chemical attack.

<sup>1</sup> First etching with hot dilute sulphuric acid, followed by polishing of the parts in relief and by a second etching with a solution of nitric acid in amyl alcohol.

TABLE III.—TENSILE PROPERTIES OF SAMPLES

No.	Mark	Minimum diameter of the test-piece after breaking, mm.	Reduction of area, per cent.	Elongation at proportional limit, mm.	Maximum load, pounds	Unit load on original section, lb. per sq. in.	Maximum elongation before fracture, mm.	Appearance of the fracture	Stretched throughout its total length	Notes
1	A <sub>4</sub>	8.15	1.33	2.00	8,210	100,100	2.20	Coarsely crystalline	do	do
2	A <sub>3-1</sub>	7.75	10.70	0.75	8,210	100,100	2.80	do	do	do
3	B <sub>3</sub>	7.85	8.35	0.75	8,490	103,500	2.75	do	do	do
4	B <sub>2-1</sub>	8.00	4.83	0.95	7,840	95,800	1.75	do	do	do
5	D <sub>3</sub>	6.60	35.20	0.25	8,120	99,100	4.00	Fibrous	Contraction well localized in a neck	do
6	A <sub>2</sub>	7.20	22.93	0.25	8,490	103,500	3.05	do	do	do
7	B <sub>2</sub>	7.00	27.14	0.25	8,850	108,100	3.90	do	do	do
8	B <sub>2-1</sub>	6.60	35.20	0.25	9,230	112,500	3.80	do	do	do
9	D <sub>4</sub>	7.05	26.16	0.40	8,490	103,500	3.25	do	do	do
10	D <sub>3-1</sub>	7.05	26.16	0.35	9,220	112,500	3.45	do	do	do
11	D <sub>4-1</sub>	6.90	29.21	0.40	8,300	101,300	3.55	do	do	do
12	C <sub>1-1</sub>	7.15	23.98	0.40	8,760	106,900	3.05	do	do	do
13	C <sub>1</sub>	5.90	48.23	0.40	7,840	95,800	4.75	Finely fibrous	do	do
14	A <sub>2-1</sub>	5.85	49.10	0.40	8,500	103,700	4.60	do	do	do
15	C <sub>2</sub>	5.85	49.10	0.50	7,750	94,600	4.10	Medium fibrous	do	do

mm.  $\times$  0.03937 = inches

124. Having thus completed a typical series showing the results of superposition of the two crystalline systems, each sample showing clearly the characteristics of both systems, let us examine the corre-

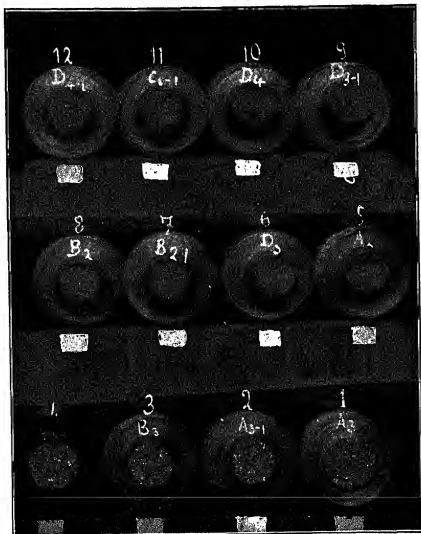


FIG. 70.—Fractures of tension test-pieces subjected to simple heat treatments.



FIG. 71.—Fractures of tension test-pieces subjected to multiple heat treatments.

sponding mechanical properties. Table III contains the results of a series of tensile tests made upon the fifteen test-pieces, treated as indicated



The test-pieces were of Frémont's cylindrical type with conical ends. In such test-piece the constricted region in which deformation previous to fracture is confined is a cylinder 10 mm. long and 8.20 mm. in diameter.

Figures 70 and 71 reproduce at natural size the appearance of the fifteen breaks. It is necessary to note that the correlation between data of Table III and the appearances shown in the last two figures, is made by the conventional marks listed in the second column of the tables, because the numbers written serially upon the half-tones do not correspond to those of the first column of the table.



FIG. 72.—Fractures after slow bending.

Table IV contains the results of some slow-bending tests made upon the first six test-pieces. These tests were made upon prismatic bars 10

TABLE IV.—FLEXURE TESTS

No.	Mark	Total maximum load, lb.	Deformation corresponding to the maximum load, mm.	Appearance of fracture	Observations
1	$A_3$	6,370	2.80	Crystalline	Test-piece stretched throughout.
2	$A_{3-1}$	7,000	3.00	Crystalline	Test-piece stretched throughout.
3	$B_3$	7,000	4.55	Crystalline	Test-piece stretched only near the fracture.
4	$B_{3-1}$	6,460	4.70	Crystalline	Test-piece stretched only near the frac-

mm. square in section, and 60 mm. long. Figure 72 shows the appearance of the six bars after failure.

Finally, Table V contains the results of a series of impact tests made upon all the fifteen treated test-pieces. These tests were made by means of a Frémont drop hammer using bars  $8 \times 10 \times 80$  mm. with a semi-cylindrical notch (2 mm. radius) across the 8-mm. face.

TABLE V.—IMPACT TESTS

No.	Mark	Total energy absorbed	Appearance of the surface of fracture
1	$A_3$	12	Coarse crystalline
2	$A_{3-1}$	18	Coarse crystalline
3	$B_3$	20	Coarse crystalline
4	$B_{3-1}$	20	Coarse crystalline
5	$D_3$	12	Crystalline
6	$A_2$	12	Crystalline
7	$B_2$	13	Crystalline
8	$B_{2-1}$	13	Crystalline
9	$D_4$	31	Fine crystalline
10	$D_{3-1}$	32	Fine crystalline
11	$D_{4-1}$	30	Fine crystalline
12	$C_{1-1}$	30	Fine crystalline
13	$C_1$	32	Short fibrous
14	$A_{2-1}$	32	Fine, long fibrous
15	$C_2$	32	Fine, long fibrous

Figure 73 shows the appearance of the fractures at about natural size.

**125.** It is first of all necessary to observe that the gradually increasing intensity of the homogeneity heat treatments to which these test-pieces were subjected, may be considered as covering a substantially complete scale, both as concerns the constituent treatments—starting from a short reheating at a temperature below  $A_{c1}$  and gradually approaching the extremely energetic treatment to which the last was subjected—and also as concerns the effects of such treatments upon the structural properties of steel.

In fact:

*First*, as far as the dendritic system is concerned, we pass from the case where it is unaffected, to the case where it is entirely obliterated by the heat treatment. See for instance the last test-piece illustrated in Fig. 66.

*Second*, as far as the network of ferrite is concerned, we also pass from the cases in which the heat treatment leaves it completely unaltered, to the cases in which it has undergone the greatest changes obtainable in practice.

*Third*, and finally, as far as the surface of the fractured metal is concerned, we also find for each physical test an absolutely complete series

called "granular," and reaching materials of a finely fibrous structure. To such structure is attributed, and often with reason, a very great importance in metallurgical practice.

We may therefore find in this series of experimental observations all the data regarding the relationships between structural characteristics and physical properties. It is to be understood that this statement is limited to the special steel to which the experiments summed up in the last few pages refer.



FIG. 73.—Impact fractures of a nickel-steel, variously heat treated.

**126.** As has been already mentioned, great importance is usually given in metallurgical practice to the appearance of the fracture, in forming a judgment of a steel's quality when considering its general suitability to withstand various physical tests. Thus a steel showing a fibrous fracture is considered to be endowed with good general mechanical properties, while a steel showing granular fracture is considered as being inferior. In the latter case, the quality of steel is thought to be poorer the more coarse and shiny the fracture.

Without entering into a discussion of the more or less sure scientific foundation upon which such a criterion may be based nor upon the

which may rest upon such a simple test, it is interesting to examine briefly the experimental data presented in the last ten sections as they indicate relationships between the form of the fracture and the better defined microstructure, as well as between the character of the fracture and those mechanical properties which may be tested with greater precision and expressed with numerical values.

**127.** A first conclusion which may be inferred from the test results on this special steel is the fact that for medium nickel-steel the so-called granular fracture is certainly not due in preponderant measure to the state of the dendritic system, but is instead very closely related to the state of the ferrite network. This is self-evident when comparing the series of structures represented by the photographs reproduced in the previous pages (see especially Sec. 119, paragraphs 2 and 3) and the corresponding data contained in Tables III, IV and V.

From such a comparison it also appears very clearly that the relation between the ferrite network and the fracture is quite different in case the fracture is caused by a gradual load or produced by impact. In fact, under gradual loading (Tables III and IV) this steel presents a fibrous fracture as early as test-piece No. 5; namely, as soon as heat treatment has obliterated ingotism—i.e. the wide ferrite network characteristic of the raw metal. From this point on, the appearance of the fracture in tension remains about constant, presenting but slight variation, except after a very energetic homogeneity treatment such as imposed upon test-pieces No. 13, 14 and 15. However, in the impact test, the granular crystalline fracture persists in all the test-pieces which were simply reheated, although always becoming finer step by step as the treatment is more elevated and prolonged, being replaced by the fibrous fracture only when the metal has been subjected to more complex and very energetic normalizing, such as those to which pieces No. 13, 14 and 15 were subjected.

It is easy to see how these conclusions more definitely establish the limits and the characteristics of the general statement that the fibrous or the crystalline fracture of a given steel depends in very large measure upon the conditions in which the fracture has been made. Confirmatory evidence will be indicated shortly.

**128.** If the results of the impact tests be more particularly examined, it must be admitted that another factor is present in addition to the very evident effect produced upon the fracture by variations of the state of the ferrite network. This second system of structural elements has a simultaneous action with the first one although of minor intensity. Thus, the variations in the ferrite and the fracture starting from the fifth test-piece ( $D_5$ ) do not by any means proceed in step. For instance, the

Fig. 58, while the fracture of the first is a great deal coarser than that of the second (see Fig. 73).

The same observations can be repeated for many pairs.

Close examination of all those figures previously shown giving the appearance of a plane section after etching with hot sulphuric acid, would seem to exclude the hypothesis that this second contributory system might be the dendritic system. In fact, a series of such photographs makes it plain that all the test-pieces which have been examined except only the last one, ( $C_2$ ) possess a dendritic system developed by hot acid attack which has not been substantially modified, much less obliterated, by heat treatment.



FIG. 74.—Test piece No. 6, after double etching.  $\times 65$ .

However, one should not deduce merely from this fact alone that the second structural system influencing the fracture cannot be the dendritic system. Nothing so far mentioned excludes the supposition that the etching figures might be *qualitatively* unaltered in form and disposition, although the absolute amount of those local differences in concentration of the various elements in solid solution might be changed a great deal by the normalization. Granting the fact that such heterogeneities are responsible for the formation of etching figures, one cannot exclude the possibility that the heat treatments mentioned have simply attenuated by means of diffusion the characteristic differences of concentration in the various parts of each dendritic crystal. In this case it will be enough that such equalization certainly cannot cause any displacement of the primary

variation in the rate of etching. This appears to explain rationally the persistence and apparent constancy in the etching figures reproduced in the preceding pages.

It seems quite probable that this hypothesis is correct, for various reasons which cannot be analyzed here, but which are founded in part upon considerations inherent to the presence and distribution of non-metallic inclusions.

**129.** To sum up the conclusions which may be deduced from the cursory observations of the last few sections: As far as concerns the relations existing between the structural systems under consideration—and for the special case here under observation—the specific function of the network of ferrite appears to be as follows:

(a) In the 2-per cent. nickel steel examined, the position and condition of the particles comprising ferrite network is independent of the position and condition of the particles comprising the dendritic system. This remark holds whether the excess constituent occurs totally or partially as a wide network, characteristic of steel as cast, as well as in the case where the network is totally destroyed.

(b) If this hypoeutectoid steel be subjected to a series of heat treatments capable of breaking up the original ferrite network, the resulting fracture obtained either by gradual stress or by impact is most highly modified by those treatments which cause the substitution of a new network for the original reticulation characteristic of raw steel.

(c) Having once destroyed the original ferrite network of ingotism, heat treatments which cause further remarkable transformations of the new ferrite system replacing the first one, produce no great variation in the fractured surfaces obtained by gradual stressing. It is necessary to have recourse to very energetic homogeneity treatments in order to obtain modifications of industrial value in the appearance of such test-pieces.

The practical importance of this observation when studying heat treatment of steel castings of this type may be easily understood.

(d) The fractures obtained by impact also vary a great deal when heat treatment destroys the original wide network of ferrite. Further modification in ferrite structure also is reflected in the fracture under impact. But in such cases variations in the fracture do not consist with variations in the ferrite, but rather appear to be a consequence of the added effect of another system of structural elements, which seems to be that of the dendrites.

**130.** Passing now to the examination of the effects which homogeneity heat treatments produce upon the most important among the physical properties of the steel under observation, which can be determined quantitatively and expressed numerically, let us examine briefly each one of

Start with the tension tests listed in Table III, Sec. 124.

As far as the ultimate strength is concerned, it apparently oscillates between 94,500 and 112,500 lb. per square inch, without its variations presenting any apparent relation to changes in the heat treatment.

This remark is fully confirmed by the examination of the Brinell hardness numbers, reported in the last column of Table I and the seventh column of Table II.

However, the other properties show more regular variations, fully confirming the conclusions arrived at already in previous paragraphs.

Thus, the first four test-pieces—in which the heat treatment has not destroyed the original *wide network of ferrite* characteristic of cast steel—present all the well-known physical properties of metals formed by two structural constituents endowed with different plasticity, and built up of crystals of large dimension. The specific mechanical function of the ferrite network is well defined by said physical properties. As we have already seen in this case and more particularly in Sec. 118, the effect of the ferrite network still predominates by far over that of any other structural system. The same physical properties also give a well-defined significance to the granular and crystalline fracture of the first four test-pieces.

The before-mentioned characteristics of the fracture persist through all the successive test-pieces which were *simply reheated*. However, reduction in area does not vary materially, and its fluctuations do not follow that of the ferrite network in any way.

Instead, the reduction of area increases a great deal in the last three test-pieces. When the fact that the variation in the condition of the ferrite is very small in the last three test-pieces, when this fact is considered in conjunction with the others above mentioned, the intervention of modifications caused by the heat treatments upon another and different structural system is clearly indicated.

Still better defined are the indications furnished by the analysis of the figures showing the elongation undergone when the metal has reached the first point of inflection in the stress-strain curves. This figure is the greatest for the first four test-pieces, which were strongly stretched on all their length, yet it falls immediately to very low values as soon as heat treatment has destroyed the ingot structure. From this point on, the materials are fibrous and have a well localized contraction and the elongation increases step by step as the homogeneity heat treatment becomes more energetic; but as in the previous cases there cannot be observed any

responding change in existing variations in the ferrite network being noticeable.

This last fact, together with the others just mentioned, seems to indicate clearly the intervention of another structural system in addition to that represented by the ferrite network.

Finally, the maximum elongation shown by the test-pieces before fractures gives similar results. In fact, it is small for the first four test-pieces, and increases noticeably for the succeeding eight (also *simply reheated*) but without following in any way the variations in the ferrite. The maximum elongation before failure is still a little greater for test-pieces Nos. 13, 14 and 15, quenched and reheated. Therefore, the previous conclusions are also confirmed in this case, especially those which have been indicated regarding variations in the reduction of area.

**131.** Examination of Table IV giving the results of static bending tests (Sec. 124) again illustrates the findings that variations in the physical properties are related to variations produced by the heat treatment in the state of *two* structural systems. The total loads at failure do not vary a great deal, while the maximum deflection increases very much as the heat treatments become more efficacious in the elimination of ingotism. In this case also, the test-pieces containing wide network of ferrite are exactly the ones which deform by stretching along a considerable part of their length.

Finally, in the impact tests (see Table V, Sec 124) the wide difference is no longer found between the physical properties of the test-pieces in which the large network of ferrite remains (the first four test-pieces of Table V) and those of the other test-pieces in which heat treatment has substituted a finer one.

This fact—which finds its complete counterpart in the variations noted in the fracture of the metal—constitutes an important new presumption in favor of the hypothesis already clearly indicated by the previous observations covering the intervention of another structural system than ferrite.

This last remark is supported by the fact that starting with test-piece No. 9 ( $D_4$ ), the first one heated at 900°C., a strong increase in the resistance to impact appears, and this without any corresponding substantial variation in the network of ferrite. From this point on the resistance to impact remains about constant although substantial variations of the ferrite network still occur in the successive test-pieces.

We have already seen in Sec. 130 that the same fact occurs in relation to the figures for elongation at the elastic limit. In that case also, sudden variations were noticed starting from the first test-piece heated to 900°C. The practical importance of such a fact is evident.

**132.** Such considerations allow one to understand more exactly the



as well as how these relationships are affected by a modification of the steel's structure upon normalizing.

In fact, it is quite natural from what has been seen in Sec. 118 and subsequently, that the habitual path of plastic deformation in the steel under consideration, when the large network of ferrite has not been destroyed, is very much different from that which the same steel shows when this network is obliterated and its filaments are replaced by a fine and compact conglomerate. In the former case the reticular structure entirely subdivides the metallic mass along surfaces occupied by a more easily deformable system than the other superposed structural system. These webs constitute in their entirety the total resisting section and constitute a *continuous* plastic system. In the latter case, the small plastic elements are isolated and therefore no longer support the totality of stress to which the metal is subjected.

In the first case it is clear that the average resistance of the material is determined by that of the ferrite constituting the large network. Observations demonstrate this to be the constituent of minor strength. The material will also necessarily exhibit the characteristic behaviour of a metal whose structure is a conglomeration of cellular nuclei bound together with a continuous plastic cement; a supposition which all the observations clearly confirm.

Once, then, the heat treatment has destroyed the continuity of the wide network, the variations in the state of subdivision of the isolated plastic elements of ferrite cannot longer exercise a major action upon the average mechanical properties of the material. From this point on, therefore, the effects upon physical properties exercised by the modifications undergone by other structural systems, principally that of the dendrites, begin to be felt in greater measure, inasmuch as they are no longer disguised by the presence of wide ferrite meshwork.

It is well known that these arguments can be applied only in the cases in which the material is subjected to stresses whose intensity increases gradually and slowly. Sudden stress or "impact" profoundly modifies the specific action of the continuous network of plastic cement (in this case the wide network of ferrite), and it is no longer the plastic material constituting such network which necessarily must give way first.

Besides, it is easily seen how all the phenomena which we have considered in the last sections constitute merely a practical illustration of the reasoning developed in Part II regarding the relations between the morphological form assumed by the structural systems resulting from the  $\gamma \rightarrow \alpha$  transformation and the heterogeneity in concentration, remaining in the  $\gamma$  mixed crystals at the initial moment of such transformation.

**133.** The 2-per cent. nickel steel to which the previous observations may apply constitutes what might be called a typical instance of

the medium-hard special steels largely used in machine construction. This is the reason why I have thought it opportune to dwell a little longer upon the analysis of the properties which such a steel may assume by virtue of normalizing heat treatments, especially as connected with the structural characters corresponding to such properties.

For the same reason it may be useful in this section to sum up briefly the principal conclusions which have been deduced from the experiments, as far as that particular type of steel is concerned:

1. Two structural systems, *viz.* a system of dendrites, and a wide network of ferrite, are in evidence in the metal as cast, when revealed by the two chemical methods of etching already indicated, and are independent of each other as far as the form and the position of their individual crystallites are concerned.

2. While the homogeneity heat treatments comprising reheatings above  $A_{c3}$  (either followed by subsequent quenchings or not) make their effects felt immediately upon the ferrite network, suppressing and transforming it into finer network in the way and by virtue of the effects studied in Part II, the dendritic system does not disappear nor does it lose its initial characters except after extremely drastic homogeneity heat treatments.

3. No relation appears between the form of the respective elements assumed by the two systems after various treatments. It is timely to point out now that it might be thought that the fine network of ferrite might possess some singularities in connection with the cavities, whose images are contained in all the microphotographs reproduced in the previous pages. Direct examination of large polished surfaces of the various samples of the steel in question excludes this hypothesis. As an example, Fig. 75 shows the structure of test-piece  $A_2$  (already reproduced in Fig. 74 after the double etching) simply polished and etched with an alcoholic solution of picric acid.

This third observation, as well as the two immediately preceding, proves that the process generally admitted to explain the genesis and transformations of the ferrite system in hypoeutectoid steels, cannot be accepted in the simpler form which identifies the system of the dendrites with that of the primary austenite. It is necessary at least to admit that the crystallites revealed by the hot sulphuric acid etching are individuals comprised of colonies of crystalline particles of primary austenite.

4. Examining the behavior of the steel under the action of gradual stresses after treatment in various ways, it clearly appears that the structural system of ferrite exerts an influence which is felt in an absolutely preponderant degree as long as this system appears in the form of the wide network characteristic of ingotism. When heat treatment leads to the disappearance of this network, substituting other networks derived

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from it, the influence of one or more other systems, endowed with characteristic critical intervals, are added to that of the first.

5. These observations sometimes allow one to attribute a significance sufficiently precise from the metallographic point of view to the



FIG. 75.—Same area as Fig. 74, repolished and etched with picric acid.

two characteristic fractures, "granular" and "fibrous." Such significance may also be differentiated for the cases in which fracture occurs under action of suddenly applied stresses and for those in which it is caused by gradual loading.

## CHAPTER XVI

### CRITERIA FOR THE SELECTION OF CASTING STEELS

**134.** As has already been said in the previous section, the 2-per cent. nickel steel experimented with constitutes a typical case of the medium-hard special steels usually employed in present metallurgical practice for the construction of machine parts designed to resist intense stresses.

As we shall see better further on, steels of such a type are well adapted to these practical applications. They may be given a great tensile strength and a high elastic limit, although retaining good reduction in area, a strong elongation and a high resistance to impact, by subjecting them to a suitable heat treatment after forging or other hot-work is completed. In other words: such steels, properly treated, may possess high tenacity and good hardness, although maintaining all those properties which are indicated with the general term toughness, a word which is in much use among workmen when passing a judgment based upon the fibrous appearance of the fracture.

However, for various reasons of expediency it is not always possible to use special steels of that group.

A first series of such reasons is of a purely economic nature. Sometimes the higher cost of the special nickel steels in comparison with the cost of ordinary carbon steels, excludes the use of the former in all those cases in which the necessity of obtaining the superior mechanical properties of which they are susceptible, does not constitute so important a condition as to subordinate considerations of cost.

On the other hand, a second series of reasons which sometimes makes it unadvisable to use the special steels of the indicated group, has a more technical nature. The manufacture, fabrication, treatment and employment of these steels are a great deal more difficult and more delicate than plain carbon steels. While the best mechanical properties which the latter may possess may be obtained with relative facility and safety even by a technical personnel devoid of special ability and experience, when using alloy steel it is necessary to enlist the aid of very capable technicians, in order to obtain with certainty and consistency the superior results which those steels may furnish. And whenever such a staff is lacking or is imperfect, so that all the complex rules which must regulate the method of manufacture and treatment are not observed and strictly controlled, it may easily happen that the same steels show results actually

have been able to reach with certainty with other less expensive carbon steels.

**135.** In case the steel need not undergo hot work such as forging or rolling, but is to be used for castings, the commercial considerations just mentioned lead to the selection of very low carbon steels, containing the proportion of silicon strictly necessary to avoid the formation of too many and too large blow holes, and also a minimum percentage of manganese to avoid the strong tendency to the formation of intercrystalline cracks or checks.

Steels of this type have a relatively low cost and do not present any special difficulty in manufacture and treatment. Also (and this perhaps is the property which renders their use advisable in the more ordinary cases) even large variations in the conduct of the melting, refining, deoxidizing, pouring, teeming, cooling and heat treating operations do not produce important variations in their mechanical properties. Thus, their use is relatively safe in the sense that it is very unlikely to cause unforeseen difficulties—even if in the hands of a second-class personnel.

Naturally these steels are not susceptible of possessing mechanical properties, and especially elastic limits, as high as those which can be obtained with steels higher in carbon and manganese, therefore their use is limited to the construction of mechanical parts which will be subjected in service only to limited stresses.

In order to give a concrete example of the characters of steels forming this group, I will select the typical case of a very soft carbon steel, in which the peculiarities are accentuated.

It was made in the basic open-hearth furnace and bottom-cast in ordinary ingots about 8 in. square in section, and was of the following composition:

Carbon.....	0.08 per cent.
Manganese.....	0.52 per cent.
Silicon.....	0.07 per cent.
Sulphur.....	0.01 per cent.
Phosphorus.....	0.03 per cent.

Figure 76 reproduces at about natural size the appearance of the surface obtained by cross-cutting one of the ingots normal to its main axis at about three quarters the total height measured from the bottom. The surface was accurately polished and etched for one and a half hours with a 20 per cent. solution of sulphuric acid, maintained at 60°C.

As is evident, such attack has not revealed a dendritic structure, properly speaking, nor did such a structure appear upon the surface even after prolonging the etching for many hours. This observation confirms the correctness of the observations made in Part I regarding the behavior of the substances which are found in solid solution in the  $\gamma$

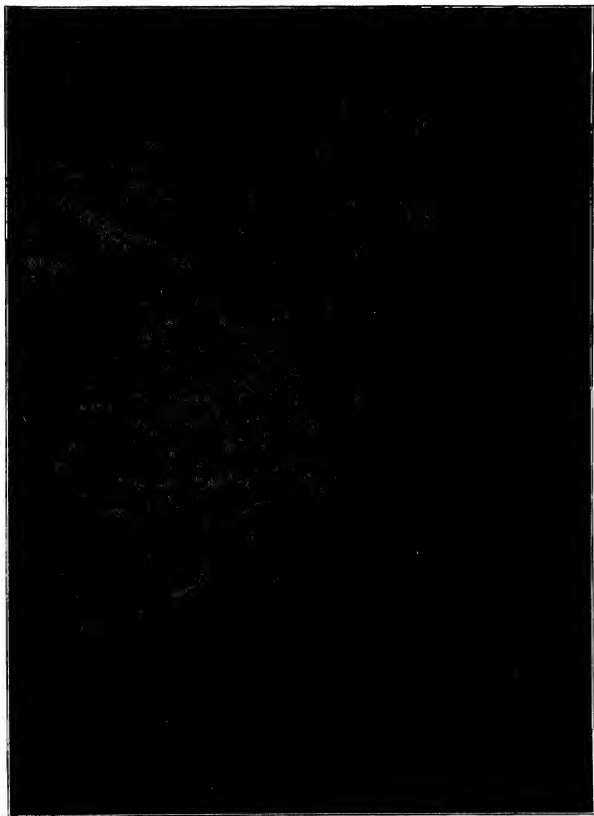


FIG. 76.—Macrostructure of 0.06-per cent. carbon steel ingot, cross cut. Etched with hot 20 per cent.  $H_2SO_4$ . Natural size.

tions in various portions of the primary dendrites. It is not pertinent to insist upon this point at this time.

What is of more interest is to examine the modifications in physical

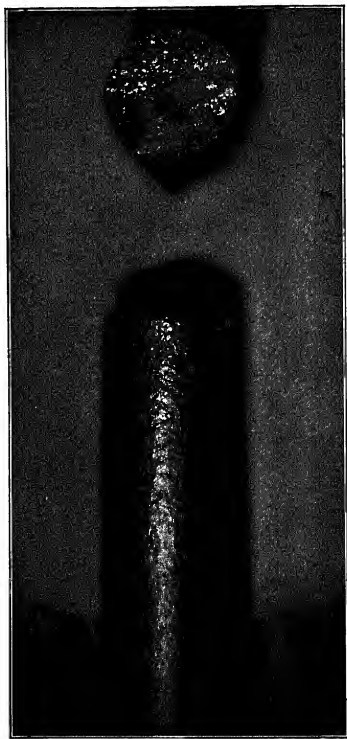


FIG. 77.—Tension test-piece of 0.06-per cent. carbon casting.

properties caused by homogeneity heat treatments of average intensity in a steel of this type in which the dendritic system is so little developed.

Two contiguous bars were cut from the piece whose section is shown

about half way the distance from center to surface.

A cylindrical tension test-piece of 13.8 mm. dia. and 50 mm. long between reference points was cut from one of the untreated bars. A test made upon this gave the following results:

Tensile strength.....	51,000 lb.
Elastic limit.....	31,000 lb.
Elongation.....	26.0 per cent.
Reduction of area.....	21.8 per cent.

The test-piece was strongly stretched throughout its entire length at the instant of failure, giving the surface a dappled appearance as can be seen in Fig. 77, showing the appearance of the cylindrical surface and the fracture.

The second test-piece was subjected to the following treatment: heated at 900°C. for four hours; quenched in cold water; reheated at 650°C. for two hours and allowed to cool slowly.



Fig. 78.—Microstructure of 0.06-per cent. carbon steel casting.  $\times 100$ .

From the bar thus treated was taken a cylindrical test-piece of the same dimensions as the previous one (13.8  $\times$  50 mm.) which gave the following results in tension.

Tensile strength.....	60,300 lb.
Elastic limit.....	42,500 lb.
Elongation.....	20.0 per cent.
Reduction of area.....	29.9 per cent.

This test-piece showed a well localized neck.



From this data it seems evident that the phenomena of ingotism are a great deal less developed in the very soft carbon steel which we have now tested than in the medium nickel steel which we examined previously. This of course could be foreseen when taking into account the known fact that those phenomena are due mainly to the wide ferrite network, which does not exist in very soft steels. In fact, ferrite constitutes the fundamental plastic mass of metal in low carbon steels, while the pearlite which fills the meshes of the ferrite network in the cast medium-hard steels forms but small islands in the former, which of themselves can not modify appreciably the continuity of the plastic mass. This fact is illustrated by Fig. 78, which shows at 100 dia. the structure of this low carbon steel as cast.

Bearing in mind the observations already made in the previous sections regarding the specific function of the wide, open network of ferrite, it would be superfluous to insist further upon this point.

As far as the numerical values of the physical properties are concerned, the data just given show clearly that the variations which may be obtained by homogeneity heat treatments are a great deal smaller in range for the very soft steels of the general analysis given than for the medium-hard nickel steel which was previously investigated.

**136.** From the point of view of steel casting practice the data which have been recorded up to this point constitute a clear confirmation of what has been already said in the previous sections on this same subject.

It is possible by the use of low carbon steels, to obtain steel castings with relative facility which possess good mechanical characteristics, and are especially free from brittleness, even when all stages of their manufacture are not supervised by exceptionally competent and experienced men. But in such cases, it is necessary to restrict the manufacture to products of low elastic limit.

However, in the manufacture of steel castings with high tenacity and particularly with a high elastic limit, it is necessary to resort to the use of medium alloy steels, and special care in the selection, execution and control of all the phases of the manufacturing process becomes indispensable in order to avoid very dangerous accidental irregularities in the quality of the products. Therefore such can be conveniently manufactured only in plants having the most perfect manufacturing equipment and control, with a thoroughly experienced organization at its disposal.

Later on I shall have occasion to mention some examples showing how it is possible to obtain consistently uniform acid open-hearth steel castings of high tensile strength and elastic limit, whose mechanical properties indicating quality either singly or as a whole are equal and even superior to those of the best forged steels. This in case the castings

opened to the application of steel products in the technique of mechanical construction.

137. Notwithstanding the evident advantage of strictly conforming to the criteria indicated in the last sections when selecting the types of steels to be used for various steel castings, it often happens that reasons of economy prevent the application of those criteria.

This is not the place to discuss whether the economy is real or apparent. It is enough to indicate that a very large proportion of the cases are based upon summary and often superficial calculations, and the result is usually given so great an importance as to overshadow the purely technical advantages mentioned in the last sections. Yet a more precise and detailed analysis of the real facts would lead to the conclusion that the economies shown by the use of low strength materials are entirely negligible and sometimes illusory. A credit is turned into a large debit if account is taken of the sometimes very great risks inherent in the use of steel castings incapable of guaranties against accidental defects and inferior to those upon which the use of better adapted materials, even if apparently more expensive, would permit a reliance with perfect confidence.

Without entering into details of this very important subject, as it would involve us too deeply in questions of cost which are out of place here, one might mention some examples to illustrate the characters of the phenomena which occur during heat treatment of the materials usually employed when it is thought best to give preponderant weight to questions of first cost.

The most frequent illustrations may be found in the manufacture of steel castings to withstand relatively high static or dead loads; in other words when they are not to be subjected to shocks, to vibrations, nor to frequently repeated forces. In a general, and admittedly rather a theoretical way, it is thought that such a casting made of plain carbon medium steel may well answer the purpose, provided its elastic limit, tensile strength and elongation have satisfactory values. The piece is then designed so that the factor of safety usually required for the special conditions considered will not be passed when the piece is withstanding the maximum combined stresses calculated to occur during use of the mechanical detail which is to be constructed. Starting from this basis and taking into consideration the fact that plain steels cost less than the special steels of a type similar to the one we have already studied, both on account of the lesser value of the raw materials entering into its making, and for some economies which can be realized in the execution of the various processes of melting, hot-working, heat treating and machining, preference is often given to the former rather than to the latter, disregarding the quality of toughness as revealed by fibrous

fractures and measured by the great reduction of area, and by the high resistance to impact. All these properties may be constantly and surely obtained in suitably treated special steels, but cannot be given a precise money value in the cost calculations for the piece to be fabricated.

The opportunity of selecting our examples in this field is the larger, since medium carbon steels are at present still widely used in the manufacture of castings for machine construction.

**138.** A first series of structural properties clearly differentiate the medium carbon steels from the medium special steels of the same type as the 2-per cent. nickel steel studied in the previous sections. Especially is this true from the point of view which interests us more particularly: that of the effects of heat treatment.

These characters are caused by the greater tendency to form that plate-like arrangement of ferrite called Widmanstätten structure. Of this we have already seen some examples in Sec. 114 and 115 when discussing a carbon steel treated in such a way as to produce the maximum degree of homogeneity practically attainable in the individual  $\gamma$  mixed crystals.

From the numerous microphotographs reproduced in the previous sections, it appears that the tendency to the formation of such structure is *nil*—or, at least, very small—in the 2-per cent. nickel steel. This character is closely connected with that other property of this steel which easily allows it to acquire a high toughness under the action of the same heat treatments.

The suppression of Widmanstätten structure is never found in equal measure in the medium carbon steels as an effect of properly selected homogeneity heat treatments. Experience shows that it is exactly to this difference of behavior in the crystalline texture to which may be ascribed the reluctance with which such steels acquire a high toughness.

Therefore, it is evidently important to examine, at least summarily, the laws regulating the formation and the development of Widmanstätten structure in the medium carbon steels.

During this rapid examination, the case which has been already indicated in Sec. 114 will not be further emphasized. In this the preliminary heat treatment has produced practically complete homogeneity of concentration of the various elements in the  $\gamma$  solid solution. The characteristics of this limiting case have been set forth in the morphological observations contained in Sec. 114, and with the general argument in Part II, Sec. 46 to 62. Instead, the only case which appears in practice will be considered. In this the heterogeneities of concentration which arose in the austenite during the primary crystallization of the steel, has only been partially reduced by later manipulation.

**139.** It is now possible to establish the following facts “a priori” on

ous observations, omitted here for the sake of brevity, confirm regarding the morphological characters of the structures resulting from the actual  $\gamma \rightarrow \alpha$  transformation:

1. If the cooling takes place extremely slowly (with a velocity of the order indicated in Sec. 114), and the crystallization of ferrite is not disturbed in any way, as it would be for instance during forging or rolling, it is clear that the first crystals of ferrite would form at the points where the carbon concentration is lowest, namely, coincident with the nuclei of the primary  $\gamma$  mixed crystals. Reasons for this statement have been minutely analyzed in Part II. Besides, the slowness with which the cooling takes place through the entire interval of temperature corresponding to the segregation of ferrite, causes those first isolated portions to exercise their maximum influence as crystallization germs, extending to all other particles of ferrite crystallizing in that neighborhood.

It follows that during quite slow cooling the crystals of ferrite start along the axes of the primary mixed crystals, and develop in branches constituting the meshes of the ferrite network. Thus originates the structure designated by Belaiew with the expression "structure of great crystals."

2. If the cooling still takes place slowly, but at a quicker rate than that indicated in Sec. 114—so that, for instance, the passage from bright red (about 800°C.) to dark (below 550°C.) takes place in a lapse of time varying between 2 and 20 hours—and the crystallization of ferrite is not disturbed mechanically, the morphological characters of the systems resulting from the  $\gamma \rightarrow \alpha$  transformation are somewhat more complex. In fact, the marshalling power of the first separated crystalline elements of  $\beta$  and  $\alpha$  ferrite cannot reach so important a value as to influence substantially all of those last particles of ferrite forming in the successive steps of the transformation. Thus a part of the last to be precipitated, not having time to migrate toward those earlier groups, orients itself around new centers of crystallization forming in interior portions of the primary  $\gamma$  mixed crystals.

Ferrite will therefore divide in this most frequent case in the following distinct portions:

(a) Ferrite of the open network, arranged in "branches" radiating from points on the axes of the primary mixed crystals.

(b) Ferrite in independent crystals developed around secondary centers of crystallization, not located at points along the axes of the primary  $\gamma$  crystals.

We have already studied in Part II, and especially in Sec. 63 to 75, the manner in which the course of these phenomena is influenced by variant concentrations of the different elements in the primary solid

it is necessary to keep the details well in mind in order to interpret correctly the crystallographic phenomena we are concerned with now.

3. Cooling takes place rapidly, though without reaching the quenching velocity necessary to prevent the  $\gamma \rightarrow \alpha$  transformation more or less completely. The marshalling action of the ferrite crystallization-germs formed at each given instant of the process upon the portions separating subsequently, now assumes smaller values the more rapid the cooling. Consequently it results that the development of the individual ferrite crystallites is reduced correspondingly, be they those forming the network or those independent of the meshes. Such cases will occur, for instance, when the steel is rapidly cooled in the air, in hot oil, or in a lead bath, as has been already indicated in Sec. 87, which processes are often incorrectly called "negative quenching." The crystalline elements of ferrite are broken up into minute fragments so that the result is a conglomeration of little crystals of ferrite, more or less uniform in size, scattered somewhat evenly throughout the kernels of pearlite. The ferrite crystals are smaller, the more rapid the cooling has been, other conditions being equal.

140. Separately examining the shape and disposition of the ferrite crystals belonging to groups (a) and (b) specified in the previous section, one might note the following facts:

1. The network appears similarly when it gathers all the free ferrite (case No. 1 of Sec. 139). In the case we are now considering however, the relative proportion of independent ferrite may increase a great deal at the corresponding expense of the network. Then the branches leading away from the primary nuclei do not grow far enough to meet each other and close the meshwork. To be exact it would then be improper to speak of a *network* of ferrite, therefore, I usually denote it by saying that we have a "network with interrupted meshes."

2. Crystals of ferrite independent of the network may appear under various aspects.

We have already seen in Sec. 66 that such crystals may form in three distinct regions of the  $\gamma$  solid solution existing on all sides of the ferrite comprising previously separated parts of the large network. Therefore, we must expect, and experience confirms this expectation, that they do not show uniform morphological characters. It would require too long a digression to analyze the various appearances caused by such diversity; it must be sufficient to say:

(a) In those austenitic regions where the carbon is below 0.35 per cent.,

Only when the precipitation of the ferrite in these regions has gone so far as to cause the carbon concentration in the neighboring austenite to rise above 0.35 per cent. does the newly separating ferrite form directly as alpha iron, taking the lamellar form which characterizes *all* the ferrite which forms in the regions to be examined in the following section, (b).

(b) Alpha iron separates directly as such in the regions where the concentration of carbon in austenite is greater than about 0.35 per cent. As has been remarked many times, alpha iron assumes a characteristic lamellar form, which appears in microscopic sections as commonly observed *needles*, more or less well developed.

Such needles are exactly those which give rise to the so-called "Widmanstätten structure" when they are highly developed.

3. As a deduction from the entire foregoing argument, it is clear that the relative division of ferrite between the network and the independent crystallites depends upon the heterogeneity of concentration in the  $\gamma$  solid solution, upon the velocity of cooling of the metal and upon the nature of the steel.

If the non-uniformity of chemical composition remaining in the austenite after normalizing is still great, although less than that degree which would cause substantially all the ferrite to become part of the meshes of the reticular structure, the amount of ferrite in the network will be larger than that existing as independent crystals in the final complex structure. The contrary ratio will obtain if heat treatment has more effectively smoothed out the original heterogeneities in the  $\gamma$  mixed crystals.

In the same way, and other conditions being equal, the proportion of ferrite in the network will be larger than in independent crystals, the slower the cooling of the system. It is particularly necessary however that the heterogeneities of chemical composition in the austenite be exactly equal in comparative specimens.

Finally, the relation between the two portions of the ferrite varies widely with the nature of the steel, always considering other conditions are equal.

Although the considerations of average chemical composition of the metal are of the highest practical importance, it is not possible to enter into a detailed discussion of them. I will content myself with pointing out a fact which, after all, could clearly be deduced from what has been seen in Part II: that the proportion of ferrite finally existing in independent crystals is higher, the greater the *frequency* of the crystallization centers forming in the austenite under given conditions of cooling. Now, it is known that this frequency is a specific character of each solid solu-

gories of steels characterized by closely spaced centers of crystallization of ferrite in the  $\gamma$  solid solution. After all one could easily infer this from the microstructure which has been examined. In fact, we have seen that numerous independent crystals of ferrite already appear in the ingot (for example see Fig. 48 and 51) in which the austenitic heterogeneity is a maximum. Also that all heatings to temperatures within the field of  $\gamma$  solid solution as demarked in the equilibrium diagram even when followed by a relatively slow cooling, gave rise to a conglomeration of finely divided ferrite crystals, in which the last trace of the large network had practically disappeared.

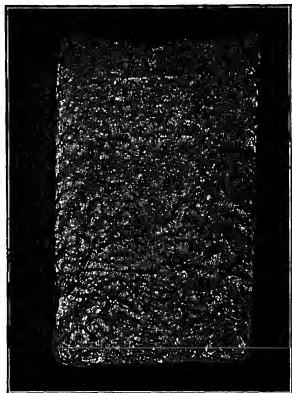


FIG. 79.—Dendritic structure of 0.33-per cent. carbon steel.  $\times 2$ . Etched with hot 20 per cent.  $H_2SO_4$ .

In order to explain the phenomena mentioned in the last three sections, better basing the illustration upon concrete examples, one might compare the experimental data given for the 2-per cent. nickel steel, with some similar data concerning another steel, characterized by a lesser frequency of crystallization germs appearing in the  $\gamma$  solid solution.

To this effect let us examine a medium carbon steel of the following composition:

polished and etched for 3 hours with a hot 20 per cent. solution of sulphuric acid in water. The dendritic structure of the metal is clearly seen although it is less prominent than in the 2-per cent. nickel steel.

Figure 80 shows the same piece at the same enlargement after being repolished and then deeply etched with a solution of 5-per cent. nitric acid in amyl alcohol. The effects of the phenomena already studied in Sec. 120 and 121 can clearly be observed here. For the reasons there



FIG. 80.—Large ferrite network of Fig. 79.  $\times 2$ . Etched with 5 per cent.  $\text{HNO}_3$  in amyl alcohol.

given the proportion of ferrite is a great deal less than the amount which should appear in a steel of the indicated composition if the  $\gamma \rightarrow \alpha$  transformation took place through a series of systems in complete equilibrium.

When reheating a number of large castings made of this steel, it happened that some of them cooled from above  $850^\circ\text{C}$ . at an extremely slow pace, requiring more than 40 hours to reach atmosphere—about  $20^\circ\text{C}$ . This happened because they were in the lower part of a pile charged upon the bottom of a very large furnace.

The structure of the steel so treated is reproduced at an enlargement of 180 dia. in Fig. 81 after polishing and etching with a solution of nitric acid in amyl alcohol. The proportion of ferrite existing as independent



crystals is clearly a great deal less than that comprising the network, which may be observed to be very broad.

A tensile test-piece, 22.6 mm. in diameter and 100 mm. long between gauge marks, taken from one of the castings so treated, gave the following results:

Tensile strength.....	83,500 lb. per square inch
Elongation.....	20 per cent.
Reduction of area.....	12 per cent.

This test-piece was strongly stretched on all its length, as seen in Fig. 82 at the right. The coarsely crystalline fracture of the same test-piece is reproduced at the right in the next figure, Fig. 83.



FIG. 81.—Broad network in 0.33-per cent. carbon steel after extremely slow cooling.  $\times 180$ . Etched with  $\text{HNO}_3$  in amyl alcohol.

A prismatic test-piece of the same steel 20 mm. square in section bent gradually and broke after a very small deflection (angle at fracture about  $140^\circ$ ). Figure 82 shows the appearance of this broken test-piece at the left, while at the left of the following figure is shown the coarsely crystalline appearance of its fracture.

Another casting of the same heat which was heated to between  $830^\circ$  and  $850^\circ\text{C.}$  and cooled less slowly than the previous, taking about 24 hours, possessed the microstructure shown at 180 dia. in Fig. 84. It is evident that the proportion of ferrite in independent crystals in com-

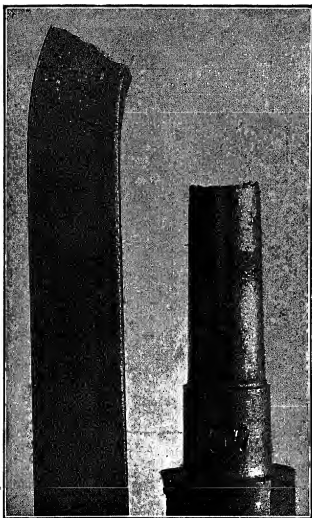


FIG. 82.—Bending and tension test-pieces, slowly cooled 0.33-per cent. carbon steel.



structure of another casting of the same steel reheated to from 830° to 850°C. for 5 hours but cooled still more rapidly than the previous one,



FIG. 84.—0.33-per cent. carbon steel cooled from 850°C. in 24 hr.  $\times 180$ .

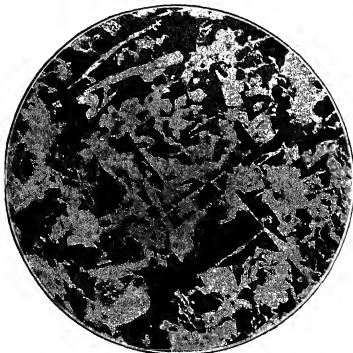


FIG. 85.—0.33-per cent. carbon steel cooled from 850°C. in 6 hr.  $\times 180$ .

*i.e.*, in about 6 hours. The large network of ferrite has almost entirely disappeared and the free iron appears in independent crystals, having the two characteristic forms indicated in Sec. 114 and paragraph 2 of Sec. 140.



FIG. 86.—Figure 85 reheated to 800°C. and cooled in air blast.  $\times 180$ .

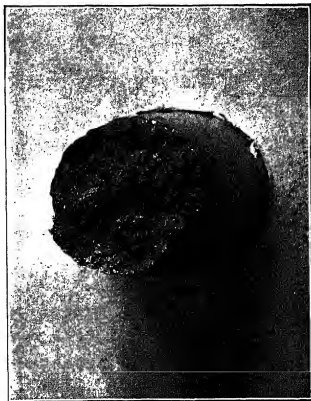


FIG. 87.—Fibrous fracture in tension fracture of 0.33-per cent. carbon steel after homogeneity heat treatment.

Finally, Fig. 86 at 180 dia. shows the microstructure of another casting of the same heat. After a first reheating and slow cooling made under the conditions just indicated for the casting whose microstructure is reproduced in Fig. 85, it was again heated to 800°C. and cooled in an air blast. Due to the dimensions of the casting, this rapid cooling did not



FIG. 88.—Bend test on 0.33-per cent. carbon steel after homogeneity heat treatment.

reach the characters of a true quenching, and allowed the transformation to take place in a substantially complete way, as can be seen clearly in Fig. 86.

This latter heat treatment is evidently that which is best for steel of this particular composition since it has fully affected the double purpose of

The microstructure thus obtained is similar to that which we have seen is obtainable by properly treating the 2-per cent. nickel steel. Experience also shows that in this case the physical properties of the two steels are comparable.

In fact, a test-piece of the steel of Fig. 86, 13.8 mm. in diameter by 50 mm. between reference points, gave the following results:

Tensile strength.....	97,300 lb. per square inch
Elastic limit.....	61,000 lb. per square inch
Elongation.....	24.5 per cent.
Reduction of area.....	36.3 per cent.

The fracture of this test was uniform and entirely fibrous as may be seen in Fig. 87. A flexure test similar to the one shown at the left in Fig. 82 bent without breaking or cracking into a U with almost parallel legs and a curvature 17 mm. in radius (Fig. 88).

**142.** It is easy to see how the observations related in the previous section are in perfect agreement with all the considerations made in Part II and in the first two chapters of Part IV. Particularly, this can be easily verified by again reading Sec. 66, 139 and 140 (§2) regarding the formation of ferrite crystals independent of the primary network; paragraph 1 of Sec. 140, concerning the formation of a network with interrupted meshes; Sec. 115 and 139 (§3) showing how heat treatment reduces the dimensions of ferrite crystals; and, finally, of the entirety of the considerations already developed.

From an industrial point of view, it may be pointed out that heat treatments of the type described at the end of the previous section, which reduce the dimensions of the ferrite crystals, are precisely those which improve within certain limits the physical properties of steels deteriorated by a reheating at too high a temperature and too prolonged a time (Sec. 115).

When compared with the results previously listed in connection with the 2-per cent. nickel steel, those given in this chapter furnish a concrete example of the quantitative limits in practice within which may be felt the tendency toward the formation of independent crystals of ferrite to the detriment of the quantity existing in the network. As has been already pointed out at the end of Sec. 140, this tendency is connected with the frequency of the crystallization centers of  $\beta$  and  $\alpha$  iron in the  $\gamma$  mixed crystals.

It is also opportune to add here another observation of the greatest practical interest:

Subdividing the ferrite into crystalline elements of small dimensions constitutes, *per se*, an obstacle to the growth of Widmanstätten structure. By its very nature we have seen that such banding presupposes a great growth in dimension of crystalline individuals of lamellar ferrite. And

accentuated, the greater the frequency of the crystallization centers of  $\beta$  and  $\alpha$  iron forming in the austenite, we must arrive at the conclusion that closely packed centers of secondary crystallization is the most important specific condition in determining the ease with which a steel responds to normalizing operations. This is the same as saying that in such a steel it is easy to eliminate the two principal causes preventing the acquisition of high tenacity, by virtue of proper homogeneity heat treatments. We have already seen these causes to consist in the tendency for ferrite to maintain the wide network of ferrite (Belaiew's "structure of great crystals") and in the tendency to the formation of large thin plates (Widmanstätten structure).

A high frequency of centers of crystallization therefore renders easy the elimination of all those characters peculiar to cast steel, which have already been studied and which are summed up by H. M. Howe in the word "ingotism." A similar tractable character is often—and not entirely improperly—denoted in practice by saying that a given steel "responds easily to the preliminary heat treatments." We have seen this to be the case of the 2-per cent. nickel steel, which yields a great deal more easily to normalizing than does the medium carbon steel examined afterwards in Sec. 141.

The tendency for ferrite to exist in wide plate-like extension has this practical consequence in a steel characterized by a great number of ferrite crystallization germs: namely, that Widmanstätten structure, with its dangerous effects upon the physical properties of the metal, develops to a great deal less extent than in a steel in which the frequency is smaller, other conditions being equal. Consequently a responsive steel—for instance, the 2-per cent. nickel steel studied—withstands the dangerous effects of overheating<sup>1</sup> which accompany a too energetic normalizing a great deal better than a steel of the inert type—such as the carbon steel studied in Sec. 141. Therefore, the former steels also show a greater factor of safety to heat treatment.

The observations made in these last pages clearly exhibit the industrial value of a correct selection of the chemical composition of the steel to be used for a certain purpose. We shall see this still better when examining other examples, to be mentioned later, not only of cast steels but of forged or rolled steels.

**143.** In order to complete the observations of this part of our study and to make sure that to the phenomena lately examined may not be attributed an absolutely exclusive value which they do not possess, one must not forget that the course of the  $\gamma \rightarrow \alpha$  transformation is strictly bound to the course of the primary crystallization, and to the distribution of various elements in the  $\gamma$  mixed crystals (in the way we have

<sup>1</sup> See Sec. 115 and 138.

studied in detail in Chap. I and II). In particular, the just mentioned effects of the frequency of ferrite germs appearing in the  $\gamma$  mixed crystals are added to those due to a larger or smaller frequency of the centers of primary crystallization which characterizes the solidification of the molten steel (and therefore to the larger or smaller dimensions of the band of variant concentration in the austenite, which is one of the most direct consequences of this frequency). As we have seen in Part I and particularly at the end of Sec. 20 and the beginning of Sec. 23, the two phenomena show concordant characters in the sense that as the centers of primary crystallization are closer spaced, the homogeneity heat treatments are rendered more rapid and more efficacious. Austenite germs thus act in precisely the same way as closely spaced ferrite germs by decreasing the distances across which diffusion must take place. Therefore, we must not forget the important fact that the frequency of the centers of primary crystallization is an element of great importance which determines the tendency of a steel to *respond* to preliminary heat treatments.

In this regard it is timely to observe that the frequency of the centers of primary crystallization is unique in that it makes itself felt *directly* upon the mechanical and physical properties of steels in which the  $\gamma \rightarrow \alpha$  transformation has been totally prevented by drastic quenching. However, the steels to which our study refers exclusively are very rarely and exceptionally used in this state. The above mentioned *direct* action has therefore a substantially lower importance than the influence it *indirectly* exercises upon the formation of ferrite, by virtue of its influence on the *dimensions of the "area of variant concentrations"* in the  $\gamma$  mixed crystals.

**144.** In this last observation, we have a more precise explanation than the one given in Part III, especially in Sec. 84 and 85, of the fundamental differences between the effects produced by simple annealings and those possible to obtain with the aid of quenching when working the soft and the medium steels ordinarily used in mechanical construction. In fact, from the theoretical considerations developed in Part II and the concrete examples noted in the last chapters, it is evident that increasing the homogeneity of the various chemical constituents in the austenite by a more or less prolonged reheating at a temperature above  $A_{c1}$  does not prevent the appearance of new and strong heterogeneities in carbon concentration on subsequent cooling—*provided it takes place with sufficient slowness*. On the contrary, we have seen how these new heterogeneities may reach very high values, even though limited to a variation in *carbon* concentration, evidencing themselves by the reappearance of the large network of ferrite or to the formation of Widmanstätten structure.

Without going over these phenomena again to examine that which



it may be useful to sum up briefly the characters by which the practical effects of preliminary annealing differ from those of preliminary quenching. This with the end in view of clarifying the exact understanding of the examples to be presented in the following pages.

A. *Preliminary annealing* is defined by the process of holding the steel at a temperature above its last point of transformation upon heating ( $A_{c3}$ ) followed by a slow enough cooling to allow the substantial completion of the  $\gamma \rightarrow \alpha$  transformation. This process has as a *constant fundamental effect* the decrease of the heterogeneities in the concentrations of the various constituent elements, except carbon. Normalization is the more pronounced the more prolonged the heating and the higher the annealing temperature. As far as carbon is concerned, it is true that we have seen that the conditions for its uniform concentration depend upon the time and temperature of the reheating, but afterwards during slow cooling this uniformity suffers such modifications as may completely alter and sometimes even reverse the effects of the previous anneal.

We have seen in Part II that this unique behavior of carbon is due to the fact that within the limits of composition comprised by the steels to which our study is restricted, only the iron:carbon solid solutions (or, more precisely, the solid solutions between iron and iron carbide) enter insoluble regions of the equilibrium diagram during cooling, corresponding to the precipitation of  $\beta$  and  $\alpha$  iron.

B. *Preliminary quenching* consists in reheating steel to a temperature above its last point of transformation on heating, followed by a rapid enough cooling to prevent the  $\gamma \rightarrow \alpha$  transformation more or less completely. It has as a *fundamental effect* the decrease of the heterogeneities of concentrations of all the elements which were found in solid solution in  $\gamma$  iron at the higher temperature reached by the treatment. Quenching is more efficacious the more prolonged and the higher the temperature of the heating and the more rapid the subsequent cooling.

We have already seen that the velocity of cooling necessary to produce the effects above indicated varies within quite wide limits with the chemical composition of the steel treated. (See especially Sec. 86 to 88.)

145. And here I may mention again a consideration of important practical interest concerning the control of the preliminary quenches by microscopic examination of the structure of the steel treated; a consideration which is founded upon observations already repeated many times during the first three parts of this volume.

We have seen that, all other conditions being equal, the diffusion of carbon in the  $\gamma$  mixed crystals takes place a great deal more rapidly than that of any of the other elements contained in steel, because of the aid given by gaseous carbon compounds. Corresponding com-

treating in many different ways.

They therefore do not appear in quenched and drawn steels. It consequently often happens in practice that a superficial microscopic examination, made only upon the steel as quenched and drawn, shows apparently a good homogeneity, when the observer has not taken the trouble to repeat it upon the same steel after having been subjected to a further true annealing at a temperature above  $A_{c3}$ . The apparent homogeneity on first examination would then be shown to hold only for carbon, and the other elements may still be distributed so discordantly as to cause such inferior qualities in the steel as to approach actual ingotism.

We have already seen very clear examples of these facts. Thus, for the 2-per cent. nickel steel which was studied somewhat in detail, it has already been pointed out that the data evidently indicates the much higher velocity with which a good uniformity of the carbon concentration was reached by means of the heat treatments, in comparison to that which approached stable equilibrium for the concentrations of the other elements. The uniform distribution of carbon shown by the state of the ferrite disguised the heterogeneities of the concentrations of other elements, heterogeneities revealed essentially by the state of the dendritic system and by the mechanical properties—particularly by the reduction of area in tension and by the impact test.

We shall see further on many other examples of the phenomena now indicated besides the one already given. However, inasmuch as the effects of such phenomena are felt in a particularly intense manner and with better defined characteristics for steels which have undergone hot-work such as forging or rolling, those examples will be reserved to the time when we shall study these last steels.

## CHAPTER XVII

### PRELIMINARY HEAT TREATMENT OF STEEL CASTINGS

**146.** The observations and experimental data referred to in the preceding chapters of Part IV illustrate with sufficient clearness the fundamental considerations developed in Parts I, II and III. Therefore it would be superfluous to repeat a detailed examination of their applications to other practical examples.

On the other hand, it is useful to show by means of some further illustrations the great variety of results possible to obtain on various types of steels in practice by applying the processes which we have studied in a theoretical outline, especially in connection with the differences existing between the effects of homogeneity quenchings and of mere homogeneity annealings. Therefore, I shall add to the data already presented, giving only some figures regarding the practical results obtained by subjecting cast steels of various types to determined heat treatments. The reader himself can make the correct interpretation of the data, after all a very easy task, if founded on the basis of criteria such as those followed in the preceding chapters.

In order not to multiply the examples unduly, some of the more characteristic and so to speak typical instances will be chosen.

**147.** A start will be made with a carbon steel for castings softer than the one examined in Sec. 141.

The chemical composition of this steel was as follows:

Carbon.....	0.23 per cent.
Manganese.....	0.66 per cent.
Silicon.....	0.21 per cent.
Phosphorus.....	0.05 per cent.
Sulphur.....	0.03 per cent.

A casting of this steel was reheated at 850° to 900°C. and allowed to cool very slowly, requiring about 36 hours to reach room temperature. It showed the microstructure reproduced in Fig. 89 (at an enlargement of 100 dia.), and the following physical properties, determined by pulling a tensile test-piece 12 mm. diameter and 40 mm. long between reference points.

Tensile strength.....	61,600 lb. per square inch
Elastic limit.....	34,300 lb. per square inch

broke at an angle of  $130^{\circ}$  with a coarsely crystalline fracture represented in Fig. 90.



FIG. 89.—Microstructure of 0.23-per cent. carbon steel; annealed at  $875^{\circ}\text{C}.$  and water cooled.  $\times 100.$

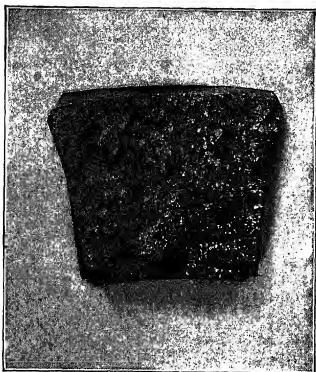


FIG. 90.—Fracture under bending. Same steel as Fig. 89.

The same steel, reheated during 12 hours at  $800^{\circ}$  to  $850^{\circ}\text{C}.$  and then cooled, requiring ten hours to reach atmospheric temperature, s

the microstructure reproduced in Fig. 91 (also enlarged 100 dia.) and the following physical properties, determined as before.

Tensile strength.....	68,000 lb. per square inch
Elastic limit.....	41,000 lb. per square inch
Elongation.....	17.0 per cent.
Reduction of area.....	13.9 per cent.



FIG. 91.—Microstructure of 0.23-per cent. carbon steel; annealed at 825°C. and slowly cooled.  $\times 100$ .

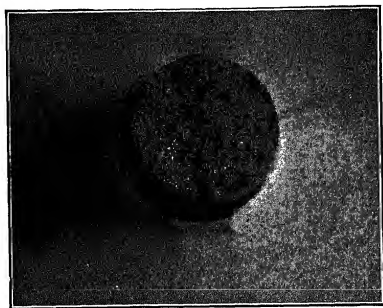


FIG. 92.—Part crystalline and part fibrous tension fracture of steel shown in Fig. 91.

The tension test revealed a mixed structure (part crystalline and part fibrous texture) as shown in Fig. 92. The static bending piece

Finally, another casting of the same steel—after undergoing the first treatment above mentioned—was reheated at 800°C. during 15 minutes, quenched in boiling water, and finally drawn during one hour at 650°C.



FIG. 93.—Microstructure of 0.23-per cent. carbon steel after complex heat treatment,  $\times 100$ .

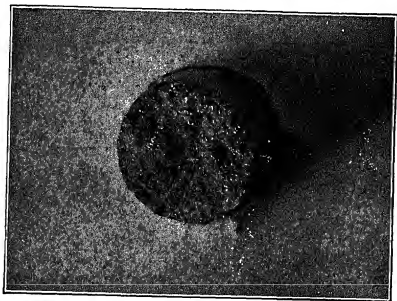


FIG. 94.—Fibrous fracture of steel shown in Fig. 93.

A test made under the identical conditions of the two previous ones gave the following results:

Tensile strength.....	73,600 lb. per square inch
Elastic limit.....	47,000 lb. per square inch
Elongation.....	

test-piece reproduced in Fig. 94 was entirely fibrous. The prismatic flexure test bent into a U similar to the previous one.

It is easy to see that all the above data are in complete accord with all the previously developed considerations, both regarding the observed structures and the test-results.

148. Another mild carbon-steel may be cited, made by the acid open-hearth process and having the following composition.

Carbon.....	0.25 per cent.
Manganese.....	0.85 per cent.
Silicon.....	0.15 per cent.
Sulphur.....	0.02 per cent.
Phosphorus.....	0.03 per cent.

It was cast in ingots  $15\frac{3}{4}$  inches (400 mm.) square. Figure 95 at 100 dia. shows the structure of this steel as cast.



FIG. 95.—Ingot structure of 0.25-per cent. carbon acid steel.  $\times 100$ .

Rectangular bars  $3\frac{1}{8} \times 5\frac{7}{8} \times 11\frac{3}{4}$  inches ( $80 \times 150 \times 300$  mm.) were taken from the lower half of those ingots, and subjected to the heat treatments indicated in the second column of Table VI. In the next four columns of the same table are noted the principal results of tension tests (13.8 mm. in diameter by 100 mm. long between reference points) taken from the bar after being subjected to the corresponding heat treatments indicated in the second column. Finally, the last

column lists the figures reproducing the microstructure of the variously treated steels after polishing and etching with an alcoholic solution of picric acid.

TABLE VI.—TESTS ON 0.25 CARBON CASTING

No.	Heat treatment	Physical properties				
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction in area, per cent.	Structure (enlarged 100 dia.)
1	Annealed at 940°C. during 45 min. followed by slow cooling in the furnace.	79,600	42,700	17.5	18	Fig. 96
2	Annealed at 850°C. during 45 min. followed by slow cooling in the furnace.	76,800	44,200	14.0	13	Fig. 97
3	Heated and cooled as No. 1, followed by a drawing at 700°C. during one hour and subsequent slow cooling in the furnace.	73,200	39,800	21.5	27	Fig. 98
4	Heated and cooled as No. 2, followed by a drawing at 700°C. during one hour and subsequent slow cooling in the furnace.	74,500	42,700	26.0	35	Fig. 99

The tabulated data are evidently in perfect accord with all the conclusions previously reached regarding the variations in structure and physical properties of soft steels, produced by mild homogeneity heatings. Note particularly what has been said relative to new heterogeneities induced by recrystallization of ferrite from a system of austenite crystals in which the greatest part of the original chemical heterogeneities has been undisturbed by insufficient annealing. Inherited non-uniformities are especially evident in the large variations in ductility after having been tempered. To be convinced on this point, compare the figures for elongation and the reduction of area of the first and second test-pieces with the third and fourth. These physical properties find perfect correspondence in the structural characters, as can be clearly seen when comparing Fig. 96 and 97 with Fig. 98 and 99.

149. In order to confirm and extend the conclusions reached in Sec.



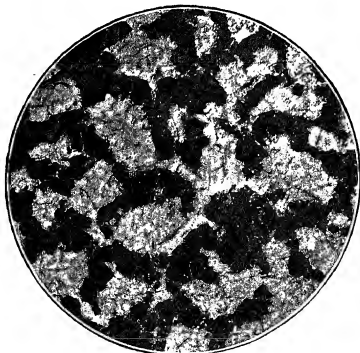


FIG. 96.—Microstructure of 0.25-per cent. carbon steel casting after heat treatment No. 1.  $\times 100$ . Etched with picric acid in alcohol.

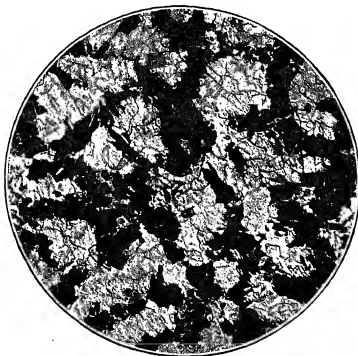


FIG. 97.—Microstructure of same casting after heat treatment No. 2.  $\times 100$ . Etched with picric acid in alcohol.



FIG. 98.—Microstructure of 0.25-per cent. carbon steel casting after heat treatment No. 3  
× 100. Etched with picric acid in alcohol.



FIG. 99.—Microstructure of same casting after heat treatment No. 4. × 100. Etched with picric acid in alcohol.

treatments produce upon the physical properties of very low carbon cast steels which, by the way, never present the real phenomena of ingotism, I may add in Table VII some comparative data between the effects produced upon the physical properties of one of those steels by a pair of homogeneity quenchings.

The steel examined was of the following composition:

Carbon.....	0.09 per cent.
Silicon.....	0.028 per cent.
Manganese.....	0.47 per cent.
Sulphur.....	0.01 per cent.
Phosphorus.....	0.02 per cent.

TABLE VII.—NORMALIZATION OF 0.09 PER CENT. CARBON STEEL

No.	Heat treatment	Physical properties			
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation per cent.	Reduction of area, per cent.
1	Heated at 850°C. for 3 hours followed immediately by quenching in water; then drawn at 650° for 2 hours.	50,500	30,000	31.8	55.8
2	Heated for 14 hours at 1100°C. followed immediately by quenching in water. Reheated at 850°C. for 3 hours followed immediately by quenching in water. Finally, drawn at 650°C. for 2 hours.	54,500	33,700	33.0	60.2

The tensile tests were made upon cylindrical bars 13.8 mm. in diameter by 100 mm. between the reference points.

After what has been said, these results do not require further comment.

When steel is cast in quite large masses, either in ingots or in formed castings, the phenomena of ingotism always appear with a great deal higher intensity than when it is cast in smaller sections. This is due to the well known fact that the slowness of cooling during the solidification range enhances the phenomena of intercrystalline liquation and segregation. Ingotism also shows a greater intensity when massive castings contain large quantities of emulsified inclusions, for the reasons indicated at the end of Part III, and especially in Sec. 98. Evidently the steel then feels the effects of even a mild homogeneity heat treatment a great deal more than it would were it cast in smaller masses.

As an example of this fact take the case of an acid open-hearth steel having the following composition:

Carbon.....	0.47	per cent.
Manganese.....	0.63	per cent.
Silicon.....	0.27	per cent.
Sulphur.....	0.029	per cent.
Phosphorus.....	0.03	per cent.

Rough castings for 381 mm. gun cradles were made with this steel weighing more than 10 tons each. One of these castings was heated for 12 hours at 750°C. Tension test-pieces (cylindrical bars 13.8 × 100 mm.) and Charpy impact samples (30 × 30 × 160 mm., with a cylindrical notch 4 mm. in diameter) were then cut from the body of this casting.

The results of these first physical tests were as follows:

Tensile strength.....	76,200 lb. per square inch
Elongation.....	19.0 per cent.
Resistance to impact (Charpy test).....	1.2 kilogrammeters per square centimeter

The pulled test-piece shown in Fig. 100 was stretched throughout all its length, and the fracture of the impact test revealed a coarsely crystal-

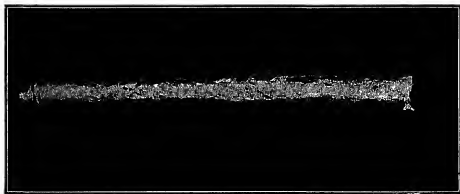


Fig. 100.—Stretched surface of tension test on a massive 0.47-per cent. carbon steel casting.

line structure, Fig. 101. These properties correspond well to the micro-structure, reproduced at 100 dia. in Fig. 102 (etched with an alcoholic solution of picric acid), and are indicative of strong ingotism partly due to the presence of a large number of inclusions. I have already spoken in Sec. 98 of the mechanism of such action by emulsified inclusions, appearing clearly in Fig. 102.

The same casting reheated for 7 hours at 850°C. and then left to cool in the air, so that its temperature dropped to 400°C. in about 3 hours, revealed the structure reproduced in Fig. 103 (also at 100 dia.).

Taking account of all the theoretical considerations presented up to this time and of the conditions in which the treatment has taken place, it can be seen that the effect of the reheating treatment upon the structure



FIG. 101.—Coarsely crystalline fracture of impact test, massive 0.47-per cent. carbon steel casting.



FIG. 102.—Microstructure of massive 0.47-per cent. carbon steel casting, revealing ingotism and inclusions.  $\times 100$ . Etched with picric acid in alcohol.

Equally remarkable are the effects upon the physical properties. In fact, the same physical tests were made upon the steel subjected to the second heating, with the following results:

Tensile strength.....	88,000 lb. per square inch
Elongation.....	24.00 per cent.
Resistance to impact (Charpy test).....	3.61 kilogrammeters per square centimeter

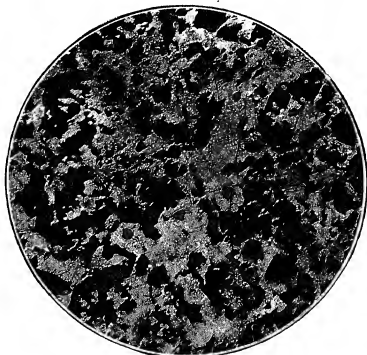


FIG. 103.—Same steel as Fig. 102 after second annealing at 850°C.  $\times 100$ .



FIG. 104.—Smooth surface of tension test piece. Steel of Fig. 100, but twice annealed.

The tension test-piece shown in Fig. 104 did not show the least stretching along its length which produces the characteristic dappled surface, and the impact fracture exhibited a finely crystalline and partly fibrous structure, Fig. 105.

This data well indicates the mechanical characteristics of ingotism appearing in thick masses of a t t el. especially when the metal contains

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In fact we must conclude that the normal properties of this steel are still altered or disguised by the effects of ingotism even after a heating at  $750^{\circ}\text{C}.$ , denoting by the term "normal properties" the average resulting from the superposition of the specific properties of its individual normal structural constituents. Were this not the case, there can be no doubt that the effect of the second heating at  $850^{\circ}\text{C}.$ , followed by slow cooling, should have changed the physical properties of steel, in an *opposite* direction to that which in reality it has produced. In fact, if the average properties of the steel would have been in large measure determined by summation of the specific properties of the individual constituents, the effect of the second heating could not have been anything but a *decrease* in tensile strength and certainly not the remarkable increase revealed by the experiment.

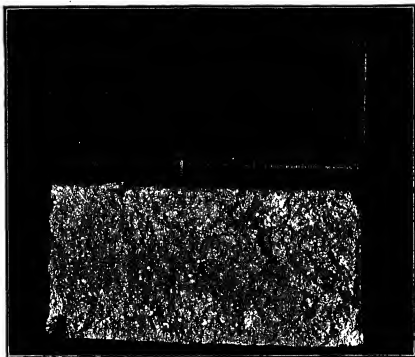


Fig. 105.—Fine crystalline and fibrous impact fracture. Steel of Fig. 101, but twice annealed.

Therefore the above example indicates clearly the specific fundamental character of all the properties constituting that special state of steel which H. M. Howe has designated with the word "ingotism." It confirms the idea that the average properties as usually determined on steel in that state are most probably due to the distribution and the continuity of the structural constituents rather than to the specific properties of the constituents themselves.

151. We should now sum up the conclusions we have reached regard-

when applied to castings for machine parts when manufactured of the types of steels more frequently used for such purposes.

In order to give examples which conform to normal industrial conditions, I shall select steels of the average purity usually attained in steel foundry operation. This observation holds not only for structural or (improperly called "metallic") impurities such as sulphur, phosphorus, copper, etc.; but also for the foreign or non-metallic impurities, especially the solid inclusions.

Results will be given for the usual physical tests, *i.e.*, tensile, static bending, and impact upon a nicked bar. In addition to these it will be useful to add the results of an impact tension test, made by the well known method using Charpy's pendulum striking a fixture holding a cylindrical test-piece 10 mm. in dia. by 50 mm. long between the conical heads. The breaking-work will be indicated in kilogram-meters necessary to rupture a 10-mm. round bar.

All the static tensile tests referred to in the following paragraphs of this chapter were made upon cylindrical test-pieces 13.8 mm. in dia. by 50 mm. between reference points.

Static bending tests were made upon rectangular prismatic bars  $9.5 \times 20 \times 160$  mm. The test-pieces were placed horizontally upon one of its 20-mm. faces, upon two parallel hard steel rolls, 70 mm. in dia., spaced 140 mm. apart on centers, and free to turn around their axes. The test-piece was forced to bend by pressure imposed upon the upper face by means of a 40 mm. steel cylinder, whose axis was placed parallel to the end supports, and so arranged that it could move up and down a vertical plane passing midway between. Pressure was maintained until the two ends of the test-piece were bent enough to slip between the two supporting cylinders, so that at the end of the test, the piece was bent like a U with almost parallel legs.

The examples to be mentioned in the following paragraphs are especially interesting when the physical properties obtainable by means of simple homogeneity reheatings and homogeneity quenchings are compared.

**152.** Starting with ordinary carbon steels for making castings, data relative to two steels containing 0.21 per cent. and 0.34 per cent. carbon are gathered in the following tables. Merely for comparison data obtained from forged test-pieces of the same steels have been added, data which will be returned to again in more detail in Part V. Working was done at about  $1000^{\circ}\text{C}$ ., hammering a square bar  $100 \times 100$  mm. down to



TABLE VIII.—STEEL "A" CONTAINING 0.21 PER CENT. CARBON

No.	Treatment	Static tensile test				Impact tensile test		Charpy impact kg.-m. per sq. cm.	Static flexure	Micro- structure	Appear- ance of the broken test-pieces
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elonga- tion, per cent.	Reduction of area, per cent.	Tensile strength, kg.-m.	Elonga- tion, per cent.				
1	Raw casting.....	75,700	42,700	25	33.5	121.0	32	4.62	good	Fig. 106	Fig. 107
2	Casting heated at 850°C. for 3 hours and cooled slowly.	77,700	53,000	27	42.0	124.5	33	8.13	good	Fig. 108	Fig. 109
3	Casting heated at 850°C., <i>quenched in air</i> and drawn at 550°C.	79,600	55,800	28	45.0	128.0	31	9.89	good	Fig. 110	Fig. 111
4	Forged bar, reheated like casting No. 2 for 3 hours at 850°C. and cooled slowly.	80,600	55,800	32	60.0	112.8	29	17.12	good	Fig. 112	

TABLE IX.—STEEL "B" CONTAINING 0.34 PER CENT. CARBON

No.	Treatment	Static tensile test				Impact tensile test		Charpy impact, kg.-m. per sq. cm.	Static flexure	Micro-structure	Appearance of the broken test-pieces
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.	Tensile strength, kg.-m.	Elongation, per cent.				
1	Raw casting.....	86,800	51,200	18	21.0	29.5	7	1.7	broke	Fig. 113	Fig. 114
2	Casting heated at 850°C. for 3 hours and cooled slowly.	85,500	54,800	25	33.5	83.3	18	6.15	good	Fig. 115	Fig. 116
3	Casting heated at 850°C., quenched in air and drawn at 550°C.	90,600	63,300	24	40.5	135.0	27	7.41	good	Fig. 117	Fig. 118
4	Forged bar, reheated like casting No. 2 for 3 hours at 850°C. and cooled slowly.	88,900	59,000	30	53.0	132.0	34	11.75	good	Fig. 119	

Carbon.....	0.21 per cent.	0.34 per cent.
Manganese.....	0.94 per cent.	0.96 per cent.
Silicon.....	0.19 per cent.	0.19 per cent.
Sulphur.....	0.03 per cent.	0.02 per cent.
Phosphorus.....	0.07 per cent.	0.06 per cent.
Copper.....	traces	traces
Arsenic.....	traces	traces

The two steels can evidently be considered as differing only in percentage of carbon inasmuch as the differences in all the other elements are within analytical accuracy.

The next to last column in the tables on pages 230 and 231 notes the figures showing microstructure of the various samples at an enlargement of 80 dia. after etching with a solution of nitric acid in amyl alcohol. The last column lists the figures reproducing the appearances of the tension and bending test-pieces. In these latter figures the two tension tests shown on the left are those broken under a gradual pull, while the two at the right are those broken under longitudinal impact. The slow bend test is shown above in the center while below is the fracture of the piece broken by the Charpy impact pendulum.

It is easy to see that all the numerical data contained in the two tables, as well as all the structures reproduced in the fourteen corresponding figures find a clear interpretation in the considerations developed in this chapter and in the preceding ones.

The general fact may again be pointed out from these two examples that the differences between the effects produced by annealing and those produced by quenching are less remarkable the lower the carbon content, all other conditions being equal. For confirmation, compare the data contained in Tables VIII and IX and the corresponding half-tones. Especially note the variations caused by each variation in the heat treatment in elongation, elastic limit, and reduction of area in the tension tests, in the tensile strength on impact and its corresponding elongations, as well as in the appearances of the fracture.

Comparing the microstructure and the physical properties of the cast sample No. 3 (rapidly cooled in air and drawn) with those of the same steel when forged and annealed (No. 4) gives a concrete example of what has been said in Sec. 145 regarding the possible errors resulting whenever it is thought that the average homogeneity of all components in a steel can be judged on the basis of a microscopical examination which, in reality, reveals only the distribution of the carbon. In the present case, for instance, the results of the physical tests reveal a greater persistency of the original chemical heterogeneities in Tests No. 3 of both steels, than in

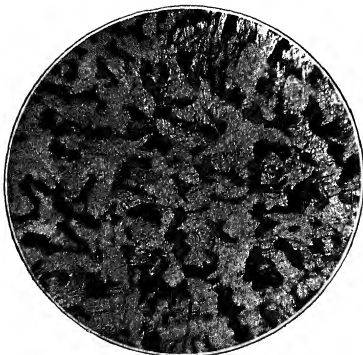


FIG. 106.—Microstructure of 0.21 carbon casting.  $\times 80$ . Etched with  $\text{HNO}_3$  in amyl alcohol.

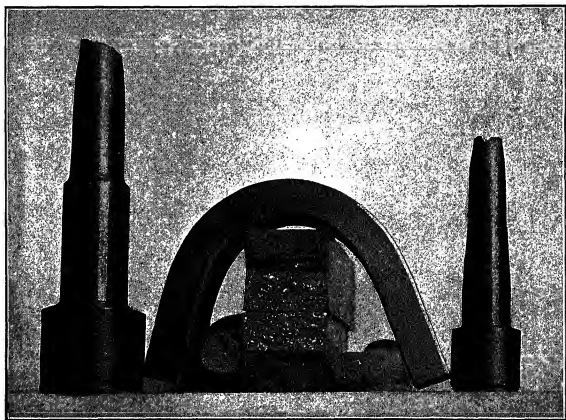


FIG. 107.—Tests on steel of Fig. 106.

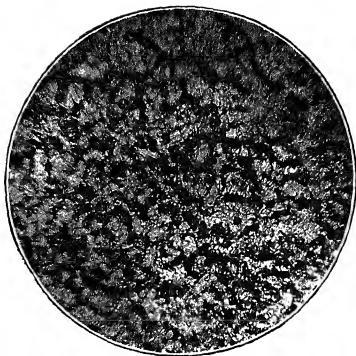


FIG. 108.—Same steel as Fig. 106 after reheating at 3 hours and slowly cooled.  $\times 80$ .  
Etched with  $\text{HNO}_3$  in amyl alcohol.

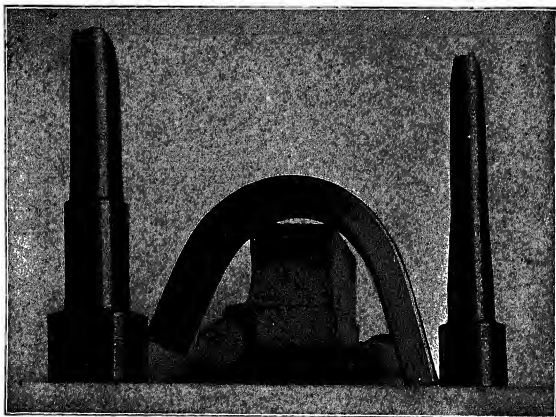


FIG. 109.—Tests on steel of Fig. 108.



FIG. 110.—Same steel as Fig. 106 after air quenching and tempering.  $\times 80$ . Etched with  $\text{HNO}_3$  in amyl alcohol.

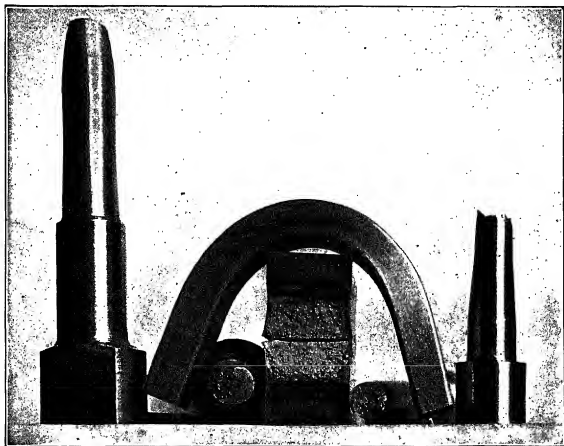


FIG. 111.—Tests on steel of Fig. 110.

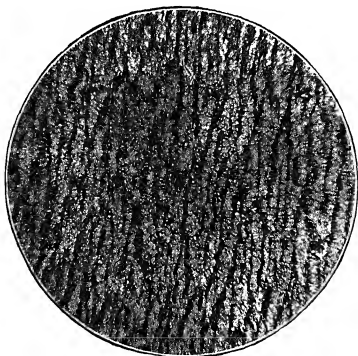


FIG. 112.—Same steel as Fig. 106. Forged and annealed at  $850^{\circ}\text{C}$ .  $\times 80$ . Etched with  $\text{HNO}_3$  in amyl alcohol.



FIG. 113.—Microstructure of 0.34 carbon casting.  $\times 80$ . Etched with  $\text{HNO}_3$  in amyl alcohol.

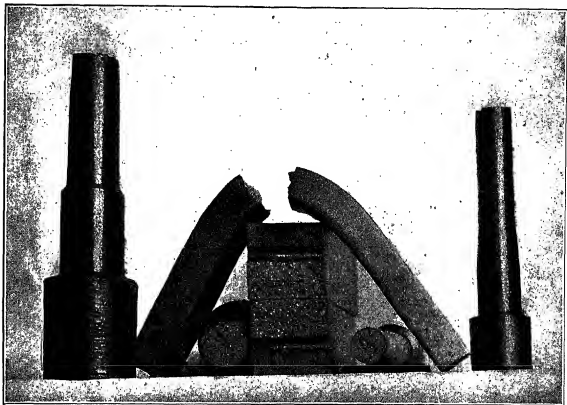


FIG. 114.—Tests on steel of Fig. 113.



FIG. 115.—Same steel as Fig. 113, after reheating at 3 hours and slowly cooled.  $\times 80$ .  
Etched with  $\text{HNO}_3$  in amyl alcohol.



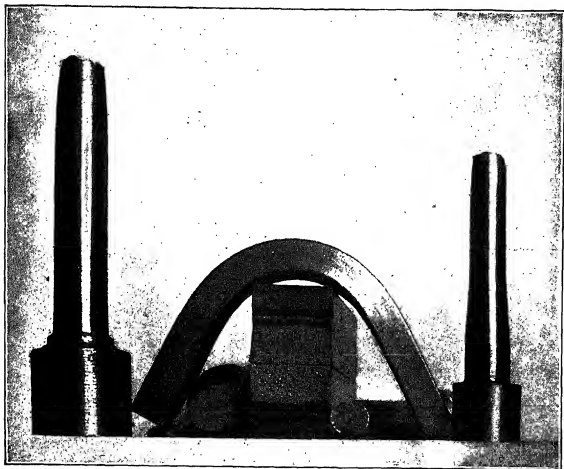


FIG. 116.—Tests on steel of Fig. 115.



FIG. 117.—Same steel as Fig. 113 after air quenching and tempering  $\times 80$  Etched

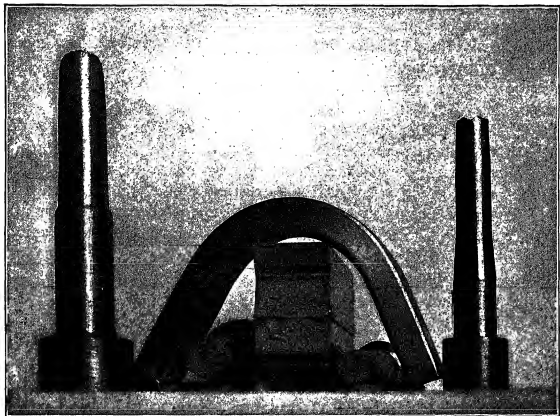


FIG. 118.—Tests on steel of Fig. 117.



FIG. 119.—Same steel as Fig. 113. Forged and annealed at  $850^{\circ}\text{C}$ .  $\times 80$ . Etched with



Twenty-three pieces like that detailed in Fig. 120 were cast from the same heat. Pieces for physical tests were taken from the thicker part of the wall of the partly square partly cylindrical bore shown in section in the lower right hand part of the drawing. The whole casting had previously been subjected to a given heat treatment.

For information and as a datum, I might add that a bar taken from an ingot of the same heat, after being forged and reheated at about 850°C. gave the following results in tension when using a cylindrical test-bar 20 mm. in dia. and 200 mm. long between reference points:

Tensile strength..... 99,500 lb. per square inch  
Elongation..... 17 per cent.

The results of the physical tests made on castings subjected to various indicated heat treatments under conditions specified in Sec. 151 are gathered together in Table X.

TABLE X.—TESTS ON NORMALIZED CASTING OF MILD NICKEL STEEL

No.	Treatment	Static tensile test				Impact tensile test		Resiliency	
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.	Tensile strength, kg.-m.	Elongation, per cent.	Kg.-m. cm <sup>2</sup> . (Charpy test-piece)	Static bending
1	Heated at 800°C. for 4 hours followed by rapid cooling in air; completely cooled in 1 hour.	96,000	56,000	15.0	14.0	106	12	5.1	good
2	Heating and cooling as No. 1; then drawn 1 hour at 700°C., followed by cooling in 6 hours.	86,800	52,500	21.0	22.5	117	27	7.5	good
3	Heated at 800°C. for 4 hours followed by a slow cooling in the furnace, requiring 18 hours.	92,200	52,000	28.5	39.8	145	35	8.22	good
4	Reheating and cooling as in No. 3 then drawn at 700°C. for 1 hour followed by slow cooling.	91,700	52,600	33.0	52.8	142	33	9.31	good

However, it is useful to point out that these data, especially when compared with those of Sec. 152, clearly substantiate a phenomenon known to be related to the relative frequency of the centers of ferrite crystallization. Special steels of this type are characterized by a closely packed or a great frequency in centers of secondary crystallization. It is therefore a great deal less dangerous to cool special steels of this type through the transformation range slowly, after a given homogeneity heat treatment, than it is to do the same thing with an ordinary carbon steel. From this special point of view it is especially interesting to compare the differences between No. 2 and No. 3 of Table X with corresponding values contained in Tables VIII and IX.

**154.** It would be easy to record a great number of other examples similar to the ones already mentioned, and illustrate still better the results which may be obtained on steel castings by means of rational normalizing. This, however, would trespass the limits within which this treatise must be kept.

Those steel castings already mentioned in Sec. 136 would be especially interesting. In them the correct selection of the steel's composition, the special precautions taken during the melting and pouring operations, and the rational application of homogeneity and final heat treatments, would procure articles possessing physical properties fully comparable to those of the best forged and treated steels.

But an explanation in great detail is at present impossible owing to the fact that they have been brought to their present perfection in very recent times, and are still held as industrial secrets only partly protected by patents. Contrary to other similar cases, secrecy is not of illusory practical worth, because the value of the processes in question depends not only upon the rules for performing one or more isolated operations, but is strictly subordinated to the rational coordination of *all* the manufacturing phases, starting from the selection of the raw materials used for making the steel and reaching to the final treatment of the fabricated piece.

It does not seem necessary to suppress some data concerning typical industrial cases, even if we cannot enter into all the manufacturing details. At any rate, it may be interesting to show the practical importance of our theoretical considerations by giving some examples of results selected from among the best which may be obtained by means of the rational application of homogeneity heat treatments to steel castings.

Above all it should be interesting to show how castings if suitably manufactured and subjected to a rational heat treatment may be substituted in many cases and with great advantage—especially as far as

rapidity of manufacture is concerned—for pieces of steel which formerly were necessarily forged and heat treated. Forged pieces naturally require very tedious and expensive machining operations in order to be brought to the final form, which in good casting practice is almost entirely obtained by the molding. Thus it has been thought interesting to select the examples just indicated from among those actual cases where special steel castings have been successfully substituted for forgings.

**155.** Even before 1915 a great deal of study was given to the fabrication and heat treatment of high-strength special steel castings at the steel works of Gio. Ansaldo & Co. obtaining results so satisfactory as to permit castings being used for a great number of gun-carriage and gun-mount parts and of machine parts which had heretofore been fabricated exclusively by tedious machining from heat-treated forgings. The use of these special castings allowed that firm largely to increase their production, and the practice has therefore been greatly extended. Experience with several thousand gun carriages under the severe and exacting use demanded during three years of the great war has fully confirmed the practical advantage and the reliability inherent in the substitution of heat-treated special steel castings for steel forgings. This substitution can fortunately be done in a great number of cases where the intensity and nature of the working stresses was such as to establish the conviction, held even by the Ansaldo Co. until a comparatively short time ago, that such substitution was practically impossible.

It is from among these very cases that the following examples will be selected.

The steels used in the Ansaldo works for the manufacture of such castings are made in the acid open-hearth, and the practice has been studied essentially with the end in view of obtaining perfect "deoxidation" of the metal, and the highest possible elimination of "emulsified" non-metallic inclusions which possess an oxidizing power upon the mass of the steel. The importance of the effects caused by such inclusions has already been mentioned at the end of Part III (see Sec. 96 to 106) and in the previous chapter (Sec. 150). Other examples will be seen further on.

Another characteristic of the steels in question, due in part to the specific action of the titanium and vanadium used in their manufacture, is the great "frequency" of the centers of alpha crystallization which form in austenite during allotropic transformation. The great practical importance of this fact has been seen from reasons advanced in the first two Parts. Numerous examples in Part IV illustrate how the potency and efficiency of the heat treatments are concerned and indicate the reliability and the uniformity of the effects produced upon such metal.

The steels we are now concerned with belong to the group of the me-

prior to any heat treatment, in fact averaging the following physical properties:<sup>1</sup>

Tensile strength.....	71,100 to 76,800 lb. per square inch
Elastic limit.....	54,000 to 61,200 lb. per square inch
Elongation.....	3 to 5 per cent.
Reduction of area.....	2 to 4 per cent.
Charpy impact test ( $\rho$ ).....	1 to 2 kg.-m. per square centimeter

A flexure test-piece breaks before even slight bending, as is shown by Fig. 121, which also reproduces the appearance of the tension test and the fracture of the Charpy impact test. The coarsely crystalline structure, typical of these steels as cast, appears still better upon the fractured surfaces obtained by bending untreated castings of medium or large dimensions in a press. Two typical examples of such structures are those reproduced in Fig. 122 and 123.

The normal microstructure of these cast steels before any heat treatment is reproduced at an enlargement of 80 dia. in Fig. 124.

Even simple reheating followed by slow cooling remarkably improves the properties of these steels. Thus annealing for about 3 hours at 850°C., imparts variable physical properties within the following limits, depending upon the composition of the steel and the dimensions of the castings.

Tensile strength.....	71,100 to 78,200 lb. per square inch
Elastic limit.....	54,000 to 61,200 lb. per square inch
Elongation.....	15 to 22 per cent.
Reduction of area.....	25 to 35 per cent.
Impact tensile strength.....	80 to 110 kg.-m.
Impact elongation.....	18 to 27 per cent.
Transverse impact (Charpy test) ( $\rho$ )..	4 to 6 kg.-m. per square centimeter.

A beam subjected to gradual loading usually breaks at an angle of about 90° to 110°, as shown in Fig. 125, which also reproduces the appearance of the other fractures. The broken surfaces still show evident traces of the dendrites.

Microstructure of the annealed steel is generally like that reproduced in Fig. 126 (enlargement 85 dia). It is easy to understand the origin of this internal structure when examining it in the light of all the considerations developed in the previous chapters.

Mild homogeneity heat treatments such as those studied in the previous chapters modify the structure which we have often seen that

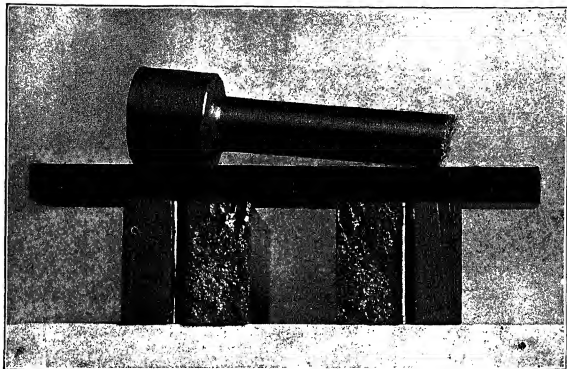


FIG. 121.—Test bars from untreated special nickel steel castings.

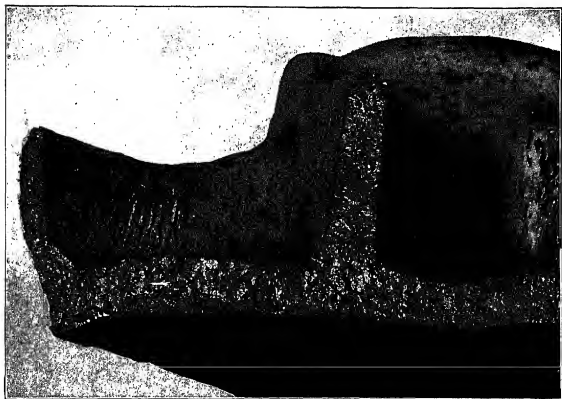


FIG. 122.—Fracture of untreated special nickel steel casting.



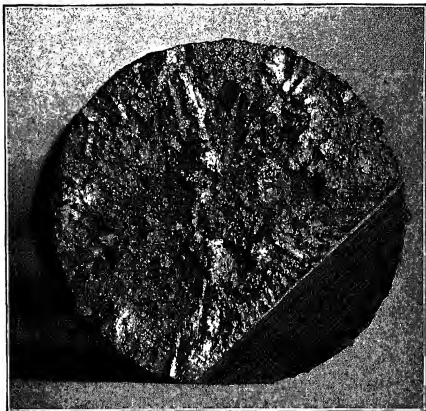


FIG. 123.—Fracture of untreated nickel steel casting.



FIG. 124.—Microstructure of untreated nickel steel casting.  $\times 80$ .

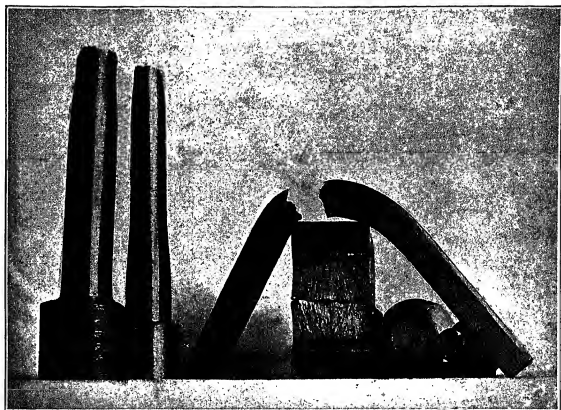


FIG. 125.—Test bars from annealed special nickel steel casting.

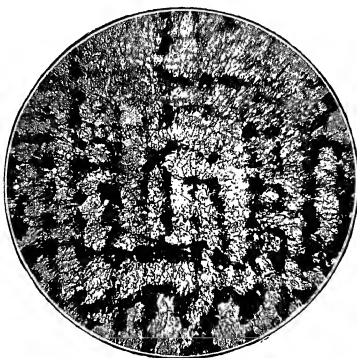


FIG. 126.—Microstructure of annealed special nickel steel casting.  $\times 85$ .

85 dia. It is superfluous to add detailed explanations of this figuration after what has already been said.



FIG. 127.—Special nickel steel casting after annealing and air quenching.  $\times 85$ .



FIG. 128.—Metal of Fig. 127 after drawing.  $\times 85$ .

A drawing operation following such an air quenching will give to the structure reproduced in Fig. 128 at 85 dia. The slight segregation is entirely normal.

structure of the same steel as the one shown in Fig. 127, when quenched in air after forging. Evidently the difference between the structures appearing under the microscope is not remarkable.

**156.** In Parts I and II we have already noted the reasons why a *simple* homogeneity heat treatment consisting of a single annealing or quenching even if followed by drawing does not give results so good as those which can be obtained by more complex treatments. Without going over arguments already discussed at sufficient length, I will only recall that this depends upon the fact that the conditions under which the various phases of a complex treatment take place may be so selected that each of them may develop one of the desired characters to a maximum degree, and up to a certain point without introducing into the

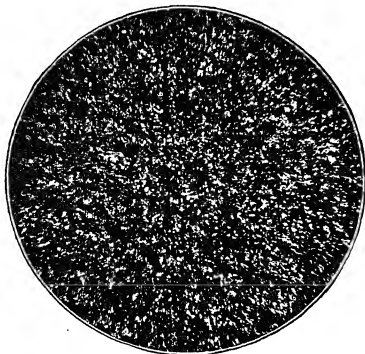


FIG. 129.—Special nickel steel casting after forging, reheating and air quenching.  $\times 85$ .

resulting steel article other dangerous defects which otherwise could eventually accompany the development of that desired property. In general such a criterion may be applied only under condition that the successive phases of a complex treatment may be conducted in such a way as to eliminate, or at least sufficiently attenuate, such dangerous effects as may take place in the previous phase.

We have seen numerous examples. Here it is sufficient to remember that even though a preliminary annealing or quenching attended by a very prolonged reheating at a high temperature may give special steels of the type now under consideration a very remarkable homogeneity of chemical composition in the austenitic state, yet on the other hand

show wide transformation intervals more or less well developed. This case is only another example added to those we have already seen, that the advantages of great uniformity in concentration obtained by means of very energetic homogeneity annealings or quenchings, cannot be conveniently exploited in industry, except under condition that further phases of the treatment may eliminate the serious injury which always accompanies the above mentioned advantages in larger or smaller measure.

It has already been pointed out that in practice the steel may be restored by means of hot work and also by means of certain heat treatments, consisting essentially of quenchings from relatively low temperatures—such as from within the transformation interval—properly conducted in such a way as to graduate their intensity. In these very cases mild quenching as in air, lead, hot oil, boiling water, etc.,<sup>1</sup> present the highest practical interest. They permit one to adjust the quenching effect at any desired intensity in the series starting from the energetic real quenchings, and ending with quenchings so prolonged as to be comparable to actual annealings. It has already been noted that the special steels lend themselves most favorably to the application of these processes. A correct selection of chemical composition permits a large variation in several different properties, the normal location of  $\gamma \rightarrow \alpha$  transformation temperatures, the metastable lowering of the limits of the extreme temperatures of said interval, and, finally, the amount of thermal hysteresis.

Very numerous combinations can evidently be obtained by applying the criteria which have been studied to determine the procedure for each step of a complex heat treatment.

**157.** The physical properties which may be obtained by applying the more complete heat treatments to the cast special steels indicated in Sec. 155, range within the following limits, according to the composition of the metal and to the dimensions of the castings treated. Compare the values given in Sec. 155 for the raw casting and after annealing.

Tensile strength.....	92,500 to 120,900 lb. per square inch
Elastic limit.....	56,900 to 96,700 lb. per square inch
Elongation.....	15 to 30 per cent.
Reduction of area.....	40 to 65 per cent.
Resistance to impact (Charpy test)...	9 to 15 kg.-m. per square centimeter
Impact tensile strength.....	115 to 145 kg.-m.
Impact elongation.....	25 to 35 per cent.

Bending tests under gradual load performed under the conditions indicated in Sec. 151 always reach the maximum deformation into a "U" without showing any crack.

Selection of steels possessing strong thermal hysteresis, relatively low transformation points and closely packed centers of ferrite crystalli-

<sup>1</sup> See Sec. 75, 76, and the first paragraphs of Part III.

zation obviates the necessity of sudden quenchings to preserve the homogeneity obtained at the high temperatures reached in one or more of the phases of the heat treatment. The main practical consequence of this fact is that the complex heat treatments may be efficaciously applied to castings of large dimensions and intricate form as well as to simple thin pieces, imparting uniform physical properties to all their parts even those of varying thickness. Moreover, the application of the desired heat treatments is safer because the elimination of quenching avoids the formation of chance quenching cracks. In the great majority

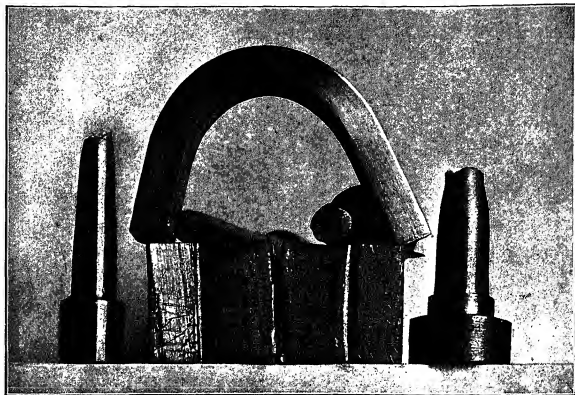


FIG. 130.—Test-pieces from heat-treated special nickel steel casting.

of cases the cooling velocities obtainable with simple air chilling is sufficient.

Figure 130 shows the appearance of the various test-pieces taken from a casting which gave the physical properties tabulated below. The bar broken by static tension is shown at the right, the Charpy impact test in the center below, impact tension at the left, and gradual bending in the top center.

Tensile strength.....	96,700	lb. per square inch
Elastic limit.....	74,000	lb. per square inch
Elongation.....	28	per cent.
Reduction of area.....	49	per cent.
Resistance to impact (Charpy test)....	14.8	kg.-m. per square centimeter
Impact tensile strength.....	122	kg.-m.



All surfaces of fracture are shown to have a finely fibrous structure. The same casting when broken by bending under a press, presents uniform fibrous fractures, from which every trace of crystalline structure has totally disappeared. For instance, Fig. 131 reproduces such a fracture of the casting from which the test-pieces shown in the previous figure were taken. It clearly shows the characteristic dull lack-luster of fibrous fractures to be expected in the broken surfaces of tough metals.

The test-piece bent like a U in Fig. 130 did not break until it was almost doubled upon itself, as shown in Fig. 132. Even then the fracture was perfectly localized, as appears better in the next Fig. 133, showing another view of the same test-piece.

**158.** Even in castings of harder steel the property of withstanding great deformations before breaking accompanies high tenacity to as great

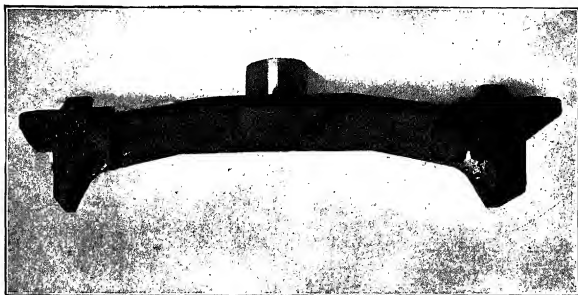


FIG. 134.—Cast, heat-treated yoke of hard nickel steel.

degree as appears from the physical properties noted in the previous section. I would like to emphasize the fact that the castings which I shall now describe were *all of high tensile strength steel*, in which the tenacity was always above 108,000 lb. per square inch and the elastic limit above 70,000 lb. per square inch. This remark is of the utmost importance in order to appraise the true import of the deformations to be mentioned shortly, deformations which could be easily obtained even after simple annealing with mild steel castings whose elastic limit would not be far from 42,000 to 50,000 lb. per square inch, but which represent exceptional results for hard high-strength steels possessing physical properties comparable only to those which can be obtained with forged and heat-treated steels.

A first example is furnished by the piece shown in Fig. 134. The necessary load to produce such deformation as shown in Fig. 135 was



single central support. After loading and bending not the least crack was found in the piece.

On increasing the pressure the piece broke partly through in the manner shown in Fig. 136. Complete fracture was obtained only when the ends were pressed together in actual contact. Figure 137 shows the surface of fracture, which was of a finely fibrous, uniform structure, characteristic of the very toughest steels.

Another cast trunnion made of high strength heat-treated steel is shown in Fig. 138. It was deformed by pressing against the two ends so as to squeeze them together in the way indicated in the next Fig. 139. During this test the high external ribs shown clearly in Fig. 140 (the same casting from the bottom) did not develop the least crack although they were badly stretched. This fact, as well as the existence of heavy stretch-

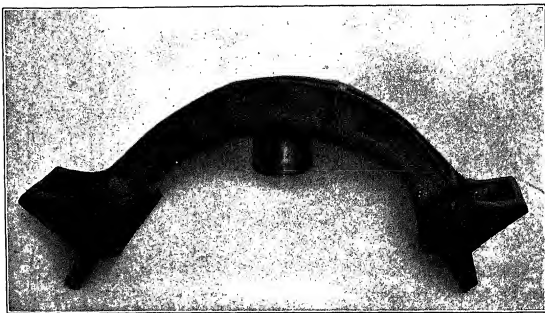


FIG. 135.—Yoke after supporting 16.5 tons at the ends.

ing in the ribs, is clearly seen in Fig. 141, viewed from the same point as the previous figure.

The sector illustrated in Fig. 142, also made of the same kind of steel, was placed in a press so that the two points of the arch were squeezed together until it was deformed as shown in Fig. 143. The metal's exceptional toughness in spite of its hardness is demonstrated especially by the absence of fractures even at places where large differences of thickness occurs, and where the lines of deformation accumulated, due to well known causes.

Finally, another large trunnion of the same kind of steel and manufactured in the same way is illustrated in Fig. 144 as it appears before cutting the two risers. It was also loaded by a hydraulic press so as to

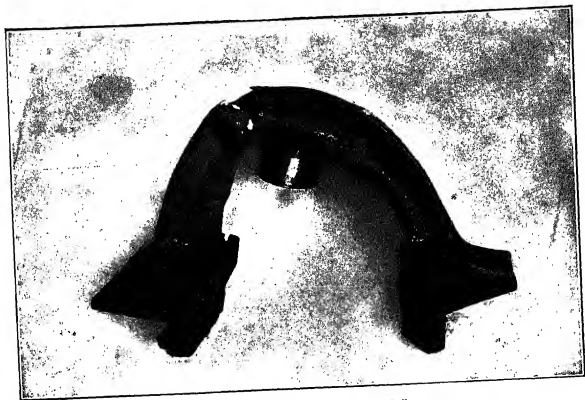


FIG. 136.—Yoke bent in press to failure.

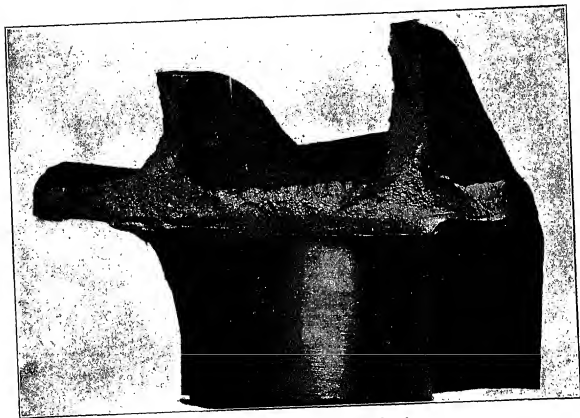


FIG. 137 —Texture of fractured yoke.

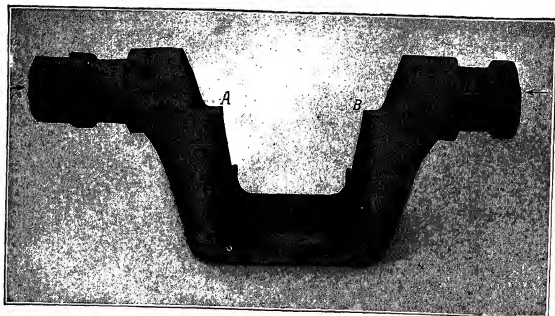
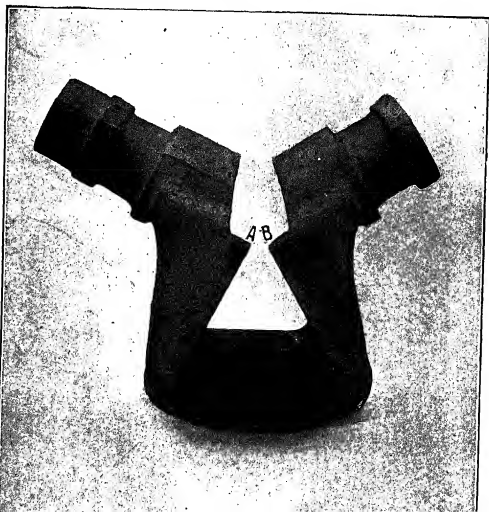


FIG. 138.—Cast trunnion of heat-treated hard nickel steel.



shown in Fig. 145, the two risers were sawed off on a hacksaw and the trunnion again placed under the press, as shown in Fig. 146. Localizing the pressure upon the trunnions by means of a block of steel, placed between the hammer and the upper trunnion as shown in the figure, the bending was continued to the point shown without causing the smallest crack in the piece.

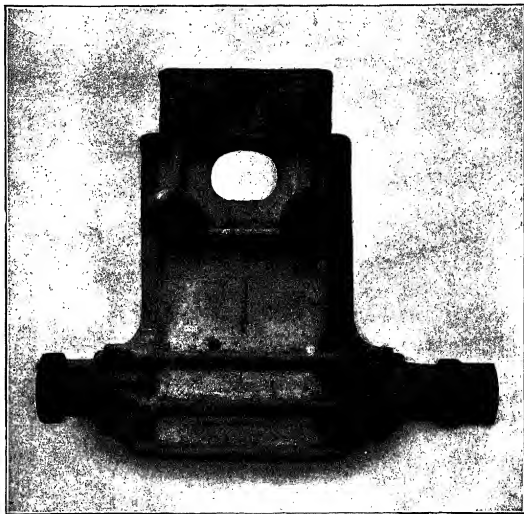


FIG. 140.—Cast trunnion from bottom; before bending.

It may be superfluous to point out explicitly that all these flexure experiments were made upon completely cold castings, at normal temperatures between  $10^{\circ}$  and  $20^{\circ}\text{C}$ .

In closing these observations regarding castings of heat-treated special steel, it may be mentioned that even during machining such castings possess toughness and plasticity equal to and even exceeding that of very mild steel castings or those of the best forged and heat-treated steels. To illustrate this fact, I reproduce in Fig. 147 at about  $\frac{1}{6}$  natural size the appearance of a turning cut from one of the hard castings in question. The figure needs no comment.

Finally, in order to show that properly heat-treated cast special steels

act in a very similar manner to forged or rolled steels when subjected to cold work, I reproduce in Fig. 148 the appearance of two test-pieces taken from two castings employed in the tests mentioned above. The two pieces indicated with the letters *A* and *B*, gave the following results:

	<i>A</i>	<i>B</i>
Tensile strength.....	123,700 lb. per square inch	101,000 lb. per square inch
Elastic limit.....	96,700 lb. per square inch	84,000 lb. per square inch
Elongation.....	21 per cent.	22 per cent.
Reduction of area.....	39 per cent.	50 per cent.

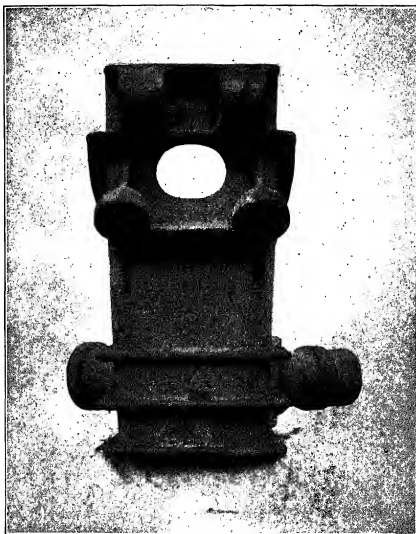


FIG. 141.—Cast trunnion from bottom, after bending.

The cylindrical part of the two test-pieces had been polished before pulling, and in their final shape show perfectly that the cold work revealed by the dull surface luster is localized in a relatively small part of the test-piece within the necked portion. It is well known that this appearance characterizes exactly those metals which have been subjected to some treatment a or olidification (similar to hot work) which has eliminated



FIG. 142.—Cast sector of heat-treated hard nickel steel.



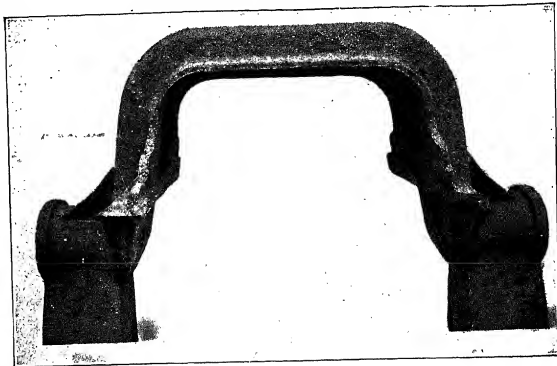


FIG. 144.—Large trunnion, a heat-treated hard nickel steel casting.



FIG. 145.—Trunnion after bending.





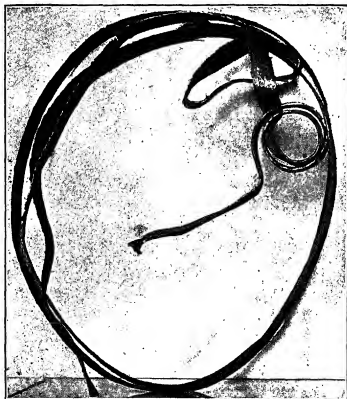


FIG. 147.—Turning from heat-treated hard nickel steel casting.

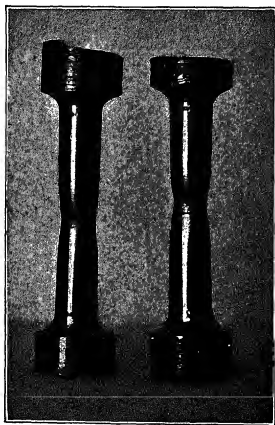


FIG. 148.—Test bars from castings illustrated in Fig. 104 to 140.

the cylindrical test-pieces are stretched over their entire length, developing ill-defined Lüder's figures throughout.

159. In the second as well as in the third and in the present part of this treatise, I have many times mentioned the fact that the phenomenon of thermal hysteresis was a very important criterion for the determination of the most suitable processes for preliminary heat treatment, and have also indicated the manner of using such a phenomenon in order to obtain

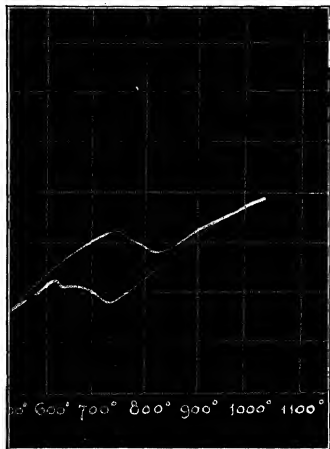


FIG. 149.—Expansion vs. temperature curves for annealed mild steel.

the best results possible with a given steel by means of said treatment. Therefore it is opportune now to add some actual data regarding the values which thermal hysteresis may assume for mild carbon steels and medium hard 2-per cent. nickel steels. These steels comprise the two principal types whose thermal reactions we have carefully studied as determining the details of the preliminary heat treatment.

Thermal hysteresis may be defined as the variation in the observed position of the transformation interval when determined upon a cooling specimen as compared to the same interval determined upon heating. Its value can be fixed experimentally by noting the rate of heat evolution

or absorption during heating or cooling at a uniform rate, particularly using the differential method devised by Le Chatelier and Saladin, or it can be found by noting the variations in the coefficient of expansion as the temperature passes the transformation interval. For the two examples to be given I have selected the second method.

Figure 149 reproduces two curves—expansion vs. temperature above and contraction vs. temperature below—characteristic of annealed mild carbon steel. The two curves are reproduced in the figure exactly as they were automatically traced upon the photographic plate in the apparatus

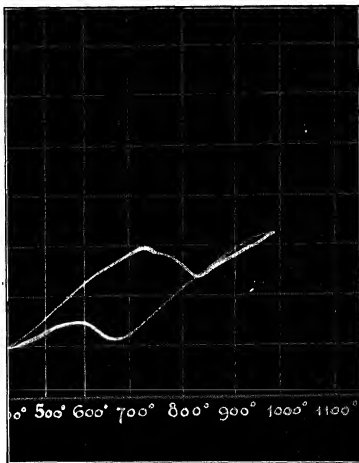


FIG. 150.—Expansion vs. temperature curves for annealed medium hard 2-per cent nickel steel.

designed by Le Chatelier and Broniewski. It would be superfluous to dwell upon the details of the tracing mechanism and the meaning of their various branches, because all this is comprised in elementary metallography.

A comparison of these curves with those reproduced in the next Fig. 150, traced under identical conditions by an annealed medium hard 2-per cent. nickel steel, clearly shows how the hysteresis interval and its

greater opportunity of selecting suitable conditions for preliminary heat treatment offered by the 2-per cent. nickel steel.

In order not to repeat things already said on this subject I will only observe that, while for the two steels the transformation upon heating ends at about the same temperature (between about  $830^{\circ}$ – $850^{\circ}\text{C}.$ ) the transformation upon cooling requires the interval from about  $740^{\circ}$  to  $620^{\circ}\text{C}.$  for the carbon steel, yet the same change in the nickel steel is starting at a temperature somewhat below  $700^{\circ}\text{C}.$  and is not entirely ended even below  $500^{\circ}\text{C}.$  In the light of all the previous remarks it is evident that the temperature interval available for genuine preliminary quenchings extends to temperatures far lower for this nickel steel than it does for the carbon steel, thus rendering these quenchings fully effective and practically applicable in many cases in which they cannot be made without great inconveniences upon carbon steel castings. The interval of temperature within which the so-called incomplete or negative quenching can be successfully noticed is also evidently the larger for such special steels.

It is necessary to remember that the dimensions and the positions of the transformation intervals, as they are located by an expansion-temperature curve similar to the ones here produced, do not have an absolute value, but, owing to the known phenomenon of lag which we have already noted several times, depend upon the velocity with which the heating and the cooling of the steel has taken place during the experiment.

The study of the changes which the dilation curves suffer with each variation in speed of heating or cooling is an important branch of the study of heat treatment of steel.

## CHAPTER XVIII

### EFFECT OF SOLID NON-METALLIC INCLUSIONS UPON PRELIMINARY HEAT TREATMENT

160. It has been my purpose to mention a series of actual examples illustrating the effects of homogeneity heat treatments on cast steels. To complete these illustrations it is useful to add another example clearly showing the magnitude of the perturbations in the course and effects of normalizing produced by the presence of certain types of non-metallic inclusions. True, in Sec. 150 I have already had occasion to



FIG. 151.—Ranked inclusions in ingot of 0.49-carbon steel.  $\times 100$ .

mention an instance where the effects of heat treatment suffered undoubted alterations from the presence of emulsified inclusions in the steel; but they were far from reaching values comparable to those which are very often reached in practice, even where it is certainly not possible to ascribe the trouble to faulty manufacture.

Therefore one of these latter cases will be mentioned at the present moment, showing the effects of emulsified inclusions when disposed in ranks,<sup>1</sup> and the improvement still possible to obtain by means of homo-

of a medium hard carbon steel made by the acid open-hearth process, and of the following analysis:

Carbon.....	0.49 per cent.
Silicon.....	0.28 per cent.
Manganese.....	0.63 per cent.
Sulphur.....	0.02 per cent.
Phosphorus.....	0.06 per cent.

The structure shown in the figure is that of the raw steel as cast in ingots weighing about 2 tons each. It clearly shows the characteristic disposition of "ranked inclusions," located in the central part of the ferrite shells (appearing as rough-bounded strips in a plane section).

Some physical tests upon the steel in this state, made under the usual conditions indicated in detail in Sec. 151, and duplicated upon the same steel after it had been subjected to various heat treatments, gave the results indicated in the following table. Beside the results of each series of tests, the table indicates the preliminary heat treatment.

No.	Heat treatment	Physical properties				Micro-structure
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Resiliency ( $\rho$ ) (Charpy test) kg.-m. per sq. cm.	
1	Cast steel.....	94,600	could not be determined	5.5	Zero	Fig. 151
2	Reheated at 800°C. during 4 hours followed by slow cooling in the furnace.	101,700	48,400	17.0	1.1	Fig. 152
3	Reheated at 900°C. during 4 hours followed by slow cooling in the furnace.	99,600	51,200	12.0	1.0	Fig. 153
4	Reheated at 900°C. during 10 minutes followed by quenching in water at 20°C. and by drawing at 600°C. during 4 hours.	122,300	83,100	1.40	1.7	Fig. 154
5	Reheated at 1100°C. during 4 hours followed by slow cooling in the furnace till 900°C. and then by sudden quenching in water at 20°C. Finally annealing at 700°C.	98,400	72,000	18.0	2.53	Fig. 155

Figures in the last column denote the illustration reproducing the structures assumed by the steel after each of the heat treatments indicated in the table (all enlarged 100 dia. after polishing and etching with an

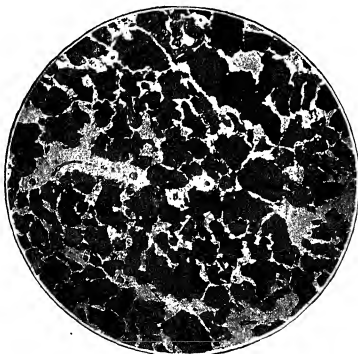


FIG. 152.—Steel of Fig. 151 after annealing at 800°C.  $\times 100$ .



FIG. 153.—Steel of Fig. 151 after annealing at 900°C.  $\times 100$ .

contained in this last table, which constitute a proof and an illustration of the propositions briefly developed at the end of Part III regarding



FIG. 154.—Steel of Fig. 151 after quenching from 900°C. and drawing.  $\times 100$ .



FIG. 155.—Steel of Fig. 151 after complex heat treatment.  $\times 100$ .

the effect which the presence of the emulsified and strongly oxidized solid non-metallic inclusions may produce upon the results of homogeneity heat treatments.



Thus, if we examine the data showing the effects of the first treatment (No. 2 of the table, and Fig. 152) one sees that the metal simply reheated at 800°C. is still very brittle, although its static tensile strength is very much improved in comparison to that of the untreated steel. In addition it is proper to observe that in this case as well as in the other example mentioned in Sec. 150, the reheating has produced an *increase* in the tensile strength, another instance that the fragility of steel before heating is to be considered as abnormal, and an indication of ingotism. Examination of Fig. 152, the microstructure of the steel after reheating for 4 hours at 800°C., shows the fact that the continuous ferrite shells developed in the unannealed steel along the ranks of inclusions are much diminished in dimension, being reduced to a few fragments gathered in the regions where the ranked inclusions are more numerous, while the remainder of the ferrite appears in the form of crystals independent of the large network. In other words, the ferrite structure has changed from the form of "large continuous network" to that which I have called "large network with interrupted meshes."

There has already been indicated in Sec. 98 the mechanism of the phenomena by which the ferrite tends to accumulate around strongly oxidized emulsified inclusions. Recalling the observations there made, it is not difficult to find an explanation of the above observed fact that an annealing prevents a full development of these continuous shells. In fact, it is sufficient to remember that diffusion always takes place during a reheating at a temperature above  $A_{c3}$ . Such diffusion of the various elements then contained in solid solution in the steel we are considering, cause their migration toward the regions occupied by the inclusions. Especially is it true of those elements whose chemical action is of a reducing nature, among which carbon predominates both because of its own high reducing properties and of its greater velocity of diffusion, due to the action of its gaseous compounds. During solidification of the ingot the oxidizing action of the inclusions had locally decreased the concentration of various constituents by oxidizing or neutralizing them. It is clear that a new influx of reducing and carburizing bodies toward the inclusions will gradually decrease the oxidizing action of the latter and, therefore, at the same time will reduce the intensity of the action which causes the accumulation of ferrite around the inclusions during the  $\gamma \rightarrow \alpha$  transformation.

We have already seen at the beginning of Sec. 106 an example of this process.

As has been briefly outlined in Sec. 98, the oxidizing action which the inclusions may exercise upon the surrounding metal is due not only to the state of oxidation of this inclusion but also depends upon the ratio which is established between the characteristic oxidation tensions

in the various steps during cooling through the transformation interval. Again, inasmuch as this last ratio depends upon a great many circumstances which are very difficult to define (among which the velocity of the cooling certainly predominates in influence) it is easily explainable how the useful effects of a long heating at a high temperature may in part be impaired or counteracted by a succeeding too-slow cooling.

An example of this last fact may be found in the effects of the second heat treatment indicated in the previous paragraph (line No. 3 of the table). In Fig. 153 after annealing at  $900^{\circ}\text{C}.$ , one may in fact see a greater development of the elements of the large network of ferrite than in the steel annealed at  $800^{\circ}\text{C}.$  (Fig. 152). Physical properties are also inferior to those obtained by annealing at  $800^{\circ}\text{C}.$ , poorer results which correspond to the microstructure of Fig. 153.

The effects of the third treatment (see No. 4 of table in Sec. 160) demonstrate that the more pronounced development of the large network observed in case No. 3 is due, as has been said, to the reformation of said network during the slow cooling of the steel, and not to the fact (which, after all, would be unexplainable) that the heating at  $900^{\circ}\text{C}.$  had altered the ingot structure less deeply than a heating at  $800^{\circ}\text{C}.$  In the third heat treatment, in fact, 10 minutes at  $900^{\circ}\text{C}.$  were sufficient to disperse the greatest part of the accumulation of ferrite existing in the large network; this could not gather together again during cooling, since quenching took place at a velocity too great to develop the processes which formerly occurred during slow cooling. We must note in passing that the steel after quenching in water was subjected to a strong drawing. The physical properties are far superior to those of No. 3, and thus confirm our conclusion.

Another proof of the conclusions reached in this section is given by the results of the fourth treatment, line No. 5 of the table. Here the homogeneity heat treatment was made under the best conditions, joining the advantages of an energetic annealing to those of a quenching, yet without falling into the known imperfections due to a quenching from too high a temperature. The structure of the steel thus treated is uniform, as shown in Fig. 155. Traces only of the large network of ferrite are retained, and the physical properties of the steel taken as a whole are better than those obtained by the previous treatments. However, the effects of the abundant inclusions are still felt, especially in the low value of resiliency ( $\rho$ ) and, probably, are due to the persistency of heterogeneities in elements other than carbon. Thus we have another example of the fact pointed out in Sec. 145 that an estimate of a steel's uniformity may be very erroneous if based exclusively upon a microscopic examina-

homogeneity heat treatments.

The following table contains all the necessary data for such comparison.

No.	Heat treatment	Physical properties				Micro-structure × 100
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation per cent.	Resiliency ( $\rho$ ) (Charpy test) kg.-m. per sq. cm.	
1	Annealed at 800°C. during 5 minutes followed by slow cooling.	97,400	48,400	12	1.1	Fig. 156
2	Annealed at 800°C. during 20 minutes followed by slow cooling.	99,300	39,800	14	1.1	Fig. 157
3	Annealed at 800°C. during 40 minutes followed by slow cooling.	99,000	37,000	12	0.9	Fig. 158
4	Annealed at 800°C. during 1 hour followed by slow cooling.	92,700	48,400	10	1.2	Fig. 159

From an inspection of these results, the effects produced by the four heatings differ very little one from the other, and the differences are irregular and do not seem to depend in any way upon the time of annealing. This is true for the variations of the physical properties as well as for the microstructure of the steel.

Regarding the irregularities in structural characters, these variations are particularly accentuated; thus, it is sufficient to note that in the steel heated during 1 hour at 800°C. (Fig. 159) the residue of the large network of ferrite is still more strongly developed than in the same steel heated at the same temperature but for only 40 minutes (Fig. 158).

It is clear that these facts constitute a further proof of the statements made in the previous section, especially the one which pertains to the reconstruction of the large network of ferrite during slow cooling in steels strongly oxidized during the melting process and containing emulsified inclusions.

**163.** Summing up the conclusions which it has been possible to reach in Parts III and IV regarding the presence of oxidized emulsified inclusions, and more especially when arranged along certain surfaces, I deem it opportune to insist upon the following points:

1. The presence of solid non-metallic inclusions of the type indicated in the last sections of Chap. XIII constitutes a severe obstacle to the regular course of those processes whose uninterrupted development is



FIG. 156.—0.49-carbon steel annealed 5 min. at 800°C.  $\times 100$ .

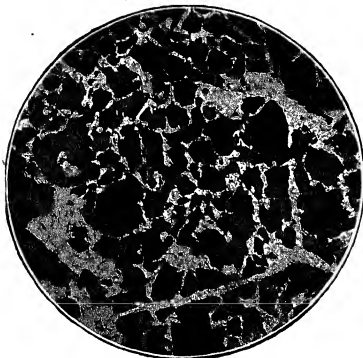


FIG. 157.—0.49-carbon steel annealed 20 min. at 800°C.  $\times 100$ .

necessary to produce all the advantages which can be obtained by means of homogeneity heat treatments.

2. From the same point of the graphic representation of the same proce

esses, as studied in the Part I and II, the effects of these sonims consist essentially in producing a relative stability and immobility in the boundaries of the band of variant concentrations existing in the  $\gamma$  solid solution.

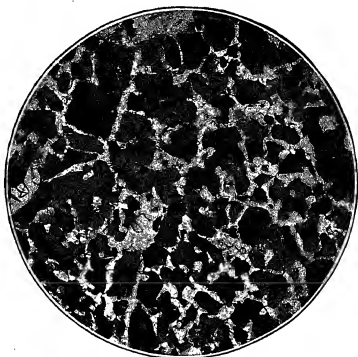


FIG. 158.—0.49-carbon steel annealed 40 min. at 800°C.  $\times 100$ .

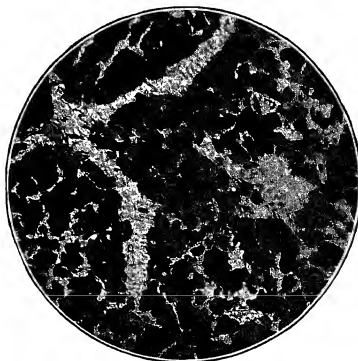


FIG. 159.—0.49-carbon steel annealed 1 hr. at 800°C.  $\times 100$ .

We have already seen particularly at the end of Sec. 63 the consequences of this last fact. Equally important effects are due to the fact that the

nature and function establishes a strict bond between the network related to the primary crystalline edifice and that of the secondary structure.

3. From the point of view of the structural characteristics of steel, the last mentioned fact is responsible for a persistent retention of the elements of the large network of ferrite, and, therefore, of ingotism.

From all the facts studied it may be concluded that unforged steels containing many oxidized sonims in ranks possess a large network of ferrite whose form and position is the result of two orders of facts; the first set of circumstances is related to the relative "distribution" of the concentrations of the various elements in the steel at the start of the  $\gamma \rightarrow \alpha$  transformation and during the change, while the second is related to the variations in the equilibrium of oxidation and reduction reactions which occur between the inclusions and the metal surrounding them during the same temperature interval.

4. From the technical standpoint, the presence of large quantities of oxidized sonims, especially if they arrange themselves *in ranks*, has a tendency to widen the known differences between the effects produced upon a given steel by a simple preliminary annealing and a veritable preliminary quenching, even when the two treatments follow the same reheating practice. It is perhaps unnecessary to remark that this is due to the fact that such inclusions facilitate the re-appearance of the elements of the large ferrite network, previously eliminated more or less completely by homogeneity annealings.

164. In the next and successive chapters we shall see how these four principal effects produced by oxidized sonims upon the properties which a given homogeneity treatment imparts to a given steel, may assume special characters and high importance should the steel have been subjected to the heat treatment after undergoing hot working such as forging, rolling, stamping, etc.

165. Before closing this chapter, it is timely again to call attention to an observation of the greatest practical importance, which is evident when all the experimental data presented in the last chapters are compared.

A great difference may thus be found in the physical and structural properties of the various steels; even the modifications which such properties suffer by reason of certain homogeneity heat treatments do not appear to be consistent even in the cases in which the chemical compositions of the steels compared and the conditions under which the respective heat treatments have been made differ very little one from the other. Even though the experimental results which we have had occasion to mention are quite limited in number, the above remark holds especially when comparison is extended to steels containing emulsified sonims more or less strongly oxidized (see, for instance, the data referred to in Sec.

gous experimental data be examined (which cannot be done here) so as to extend the comparisons to an exhaustive series of experiments in which the chemical compositions and the conditions under which the heat treatment has been made would vary in a gradual and progressive way, very strong irregularities in the corresponding variations of physical and structural properties would be easily revealed in the steels so treated.

All these facts, whose practical importance is evident, confirm and generalize the conclusion already reached in a precise way concerning some particular phenomena, such as the state of oxidation of steel, the nature of the sonims included in it, etc. I refer to the undoubted fact that the physical and structural properties of a given steel and the variations which said properties suffer by effect of predetermined heat treatments, may depend in very large measure upon many causes, which cannot always be easily determined. Among these causes, beside the others already mentioned, may be indicated as examples, the kind of raw materials used in the steel making, the refining and deoxidation of steel—whose effects sometimes are felt in a way inexplicably remarkable and persistent—the temperature and the other conditions during teeming, the form and dimensions of the ingots and castings, etc.

We shall see further on, how other irregularities and apparent anomalies accrue in steels subjected to hot working before heat-treatment.

## PART V

### PRELIMINARY HEAT TREATMENT OF FORGED AND ROLLED STEELS

#### CHAPTER XIX

#### A GENERAL DISCUSSION OF HOW THE EFFECTS OF HEAT TREATMENT ARE MODIFIED BY HOT WORK

166. The paucity and inaccuracy of the knowledge even now possessed by many mechanical constructors regarding the properties which may be imparted to cast steels by means of a rational homogeneity heat-treatment, causes them to continue to specify forged or rolled steels for the construction of machine parts which in a great many cases could be manufactured with the same and even greater safety and sometimes with a great saving in time and cost from steels of correct composition, simply cast in molds and then subjected to a properly selected heat treatment.

On the other hand, even when using forged or rolled steels, it very seldom happens that the mechanical constructor is able to avail himself fully of the advantages of homogeneity which the application of a rational preliminary heat-treatment would allow him to obtain. Consequently even in many modern machines are found parts made of forged or rolled steel blocks, and a great deal heavier than necessary. This is due only to the fact that the designer has taken into account exclusively the properties of the metal resulting from a simple hot-working process such as forging, pressing, or rolling, and completed at best merely with a simple *final* heat treatment, intended to impart a higher tensile strength than that produced by the mechanical work. Such a final treatment for strength may be a quenching, followed or not, as the case may be, by a drawing. Sometimes it is the practice simply to eliminate the internal stresses caused by mechanical work by a final drawing or annealing.

On the contrary, when the mechanical constructor and designer is able to avail himself surely and at will of the advantages obtainable in what we know as preliminary heat treatments for homogeneity, modern steel metallurgy offers him many other resources. This statement holds even in the still frequent cases in which it is not yet possible or convenient to substitute steel castings for forged, rolled, or stamped articles.

We have already examined the structural origins of these metallurgical advantages. The examples which I propose to show in this last part of



heat-treatment. It is not less important in case the steel has been previously subjected to a hot mechanical working than in the other cases which we have already studied where the homogeneity heat-treatment is applied to raw cast steels. On the contrary the practical advantages in question are sometimes greater in the former than in the latter instance. Just to mention an illustration of this remark, many examples of which shall be seen further on, it may be said that the principal purpose of homogeneity heat treatment often consists in the elimination of dangerous transverse fragility—a term I have defined as the sum of all the properties which make a specimen weak transversely—whose presence would wholly prevent the construction of pieces such as large caliber gun tubes.

Disregarding for the time being, however, the always very great and sometimes essential but purely technical advantages of the rational application of the preliminary heat treatments to forged or rolled steels, such practice may enable one in many cases to effect remarkable economies and savings (with the same factor of safety) by allowing the use of steels costing a great deal less than those steels now customarily specified to be used after a “final” heat treatment, such as indicated above.

Finally, many of the most important modern machines, as for instance, aeroplane motors, light internal combustion motors for boats, and so on, are made possible only by the rational use of special steels subjected to efficient normalizing after forging or rolling.

**167.** Usually, the nature and general course of the phenomena responsible for the effects of preliminary heat treatments are not much changed owing to the fact that this normalizing is done upon steels previously subjected to hot-work rather than upon unworked cast steels. Therefore, as far as concerns the general application of the arguments developed in Parts I and II, what I have said in Part IV remains unaltered as it refers to the examples which will be given in this and subsequent chapters.

However, the same thing cannot be said when one considers the *effects* produced upon the physical, mechanical and structural properties by a given homogeneity heat-treatment when applied to a steel of a given composition, regardless of whether it had been worked or not, because if its ingot structure had been thoroughly broken up, strong quantitative and qualitative differences will appear in the results which are then obtained when compared to those when starting with cast steels.

A general study of the source and the nature of these differences cannot be presented at this place because it would require a detailed treatise covering many groups of crystallographic facts which have no direct connection with those already studied in the previous sections, and upon

heat treated steels will become sufficiently and clearly explained by the concrete examples which will be shown in the following sections.

Therefore, in advance of these facts to be developed, I shall only mention some brief general observations, constituting simply an orderly enumeration of the principal groups of phenomena to which can be attributed the differences between the physical, mechanical and structural properties which a given preliminary heat treatment may produce in a forged or rolled steel and the corresponding properties which the same treatment produces when applied to cast steel.

We have already mentioned in Part III<sup>1</sup> the general characters of some phenomena with which we shall now be concerned, but at that time we did not have the terms of comparison, now furnished by the experimental data recorded in Part IV, necessary to complete those remarks. Therefore, it is now opportune to rehearse the argument then merely suggested.

**168.** First of all, in comparing the effects produced by a given heat-treatment in the two cases it must not be forgotten that during hot-work the steel is continuously subjected to an important homogeneity heat-treatment whose efficacy may be very great. Especially is this effective when the working, which usually starts at temperatures at which the iron is totally in the  $\gamma$  state, is so prolonged that during its last periods the steel has cooled till it reaches and partially crosses the transformation interval.

In case the mechanical work is entirely completed while the iron yet remains hot enough to be held in the  $\gamma$  modification, the effects which the resulting plastic<sup>2</sup> deformations produce upon the course and upon the results of the subsequent normalizing may be considered as resulting from the following groups of phenomena, all of which usually operate simultaneously:

1. Decrease of the original distances between points in the cast metal corresponding respectively to the maximum and minimum concentrations of the various elements contained in solid solution.

2. Variations of the orientation of the primary crystallites and of their colonies, with consequent variations of the orientation of the secondary crystalline particles resulting from the  $\gamma \rightarrow \alpha$  transformation.

These groups of phenomena exercise a remarkable *direct* action upon the course of the phenomena utilized for normalizing practice, as we shall see later by concrete examples.

3. Modifications of the crystalline edifice of the steel by the known phenomena of "slip" and "twinning," caused by plastic deformation on a metallic mass.

<sup>1</sup> See Sec. 94 and 95.

<sup>2</sup> This term is used to distinguish here between the characters of a plastic and

As is well known and has been already mentioned in Sec. 94, the phenomena of this group deeply influence the mechanical properties of steel. But the modifications so caused are in great part independent of the modifications produced upon the same physical properties by the phenomena of diffusion. Homogeneity heat-treatments are based upon diffusion alone, so that at least in good measure the effects of the two orders merely add one to the other.

It is true that in some definite cases which will be mentioned later, the effects of homogeneity heat treatments may even be manifested principally as modifications in the crystalline structure precisely similar to those mentioned in 1, 2, and 3 as depending strictly upon the previous hot work. But there is no doubt that in such cases the additional crystalline modifications taking place during normalizing are only a continuation of the same phenomena of slip and twinning in the crystalline elements, which had been initiated and, so to say, "prepared" by previous mechanical working, and then continued by the more or less prolonged heating at high temperature. At any rate, phenomena of category 3 have nothing to do with those of diffusion, which we have studied, and upon whose commercial utilization are founded the criteria for the execution of preliminary heat-treatments. Therefore these phenomena can be considered only as constituting collateral factors which accompany the diffusion processes, operating in such a way that their effects are added to those of the latter even for that part which takes place during the heating operations.

This statement naturally does not exclude the evident fact that the phenomena now indicated must be taken into careful consideration in practical heat treatment, inasmuch as the magnitude of their effects may sometimes exceed those of diffusion. But it may be repeated that these are always phenomena whose course is not practically related to diffusion. Therefore their study must be conducted separately and cannot be an integral part of the study of the processes upon which normalizing practice is founded.

A fourth group of phenomena has to be added to these three, as has been indicated above, when at least a part of the hot work takes place within the interval of temperature where the  $\gamma \rightarrow \alpha$  transformation occurs. The characteristics of this fourth group may be defined briefly in the following way:

4. Modifications of the forms, dimensions and arrangement of the  $\alpha$  and  $\beta$  iron crystals, due essentially to a "disturbed crystallization" of these structural elements, a process caused by the deformations imposed upon the metallic mass during their separation.

Let us now examine separately, even if in a summary way, the general characteristics of the phenomena pertaining to each of the above four

to 95 upon the same subject, but in the less precise form permitted by the scant ideas upon which we could then base our arguments.

**169.** In regard to the phenomena of the first group, there is nothing to add from a general point of view to what has already been said in Sec. 94. Shortening the distance between maximum and minimum concentrations will result in modifications of the processes there explained at length, modifications so simple that the explanation of their effects is so intuitive as to render unnecessary any preliminary consideration to illustrate the numerical examples which we shall have occasion to exhibit further on in this chapter. Such data alone will be enough to place in full view the soundness of the conclusion already reached in Sec. 94 concerning the greater efficacy of a given homogeneity heat-treatment when applied to a forged or rolled steel than to the same steel as cast, provided always that in judging the effects of the heat-treatment proper, the actual homogeneity-annealing which necessarily accompanies hot-work is always taken into account. In this manner a true comparison may be established with the modifications which a given heat-treatment produces upon the properties of a steel not worked in any way before being subjected to a suitable preliminary heating. Besides, when weighing the comparative data it is necessary to take into account the possible perturbations due to the intervention of phenomena of the other three groups, or to the intervention of other disturbing causes, such as, for instance, the presence of abundant and strongly oxidized sonims.

It is perhaps timely again to point out—although the observation is obvious—that the efficacy of this first group in aiding and completing the diffusion processes is the greater, the larger the primary crystalline elements and their colonies which characterize the raw steel, both on account of the chemical composition of the steel and the conditions under which it has solidified.

The effects of the phenomena of this first group, therefore, other conditions being equal except that the metal has not been subjected to any mechanical work, are especially felt with greatest intensity upon steel cast in large ingots.

**170.** The phenomena of the second group mentioned in Sec. 168 are a great deal more complex. These phenomena depend upon the variations in orientation of the primary crystalline elements and of aggregations or colonies formed by them, and upon subsequent modifications of the effects of such changed orientation which a treatment for homogeneity may cause.

We have already presented in Sec. 110 an example taken from a particular ingot showing that concordant orientation of the principal axes of the primary dendrites may occur even in certain parts of a given mass of cast steel. For obvious reasons such a condition of affairs is responsible for more or less marked anisotropy in the properties of the metal. Here-

tofore I have not thought it timely to dwell upon the characteristics of this condition because in practice it does not present sufficient importance in steels simply cast in ingots or in molded castings.

On the other hand, the analogous phenomena appearing in forged or rolled steels are of quite different weight. We shall see later that their effects often assume a by far preponderant importance.

In the greatest majority of commercial cases, the deformations produced by the mechanical work obviously result in a great increase in either one or two of the dimensions of the original billets or blanks, or at least of some portion of them. Constituent elements of the steel—primary crystals and their colonies or products of their transformation—suffer a deformation which may in some cases be considered as limited essentially to a “stretching” in only one direction. This happens, for instance, to an ingot during rolling or forging into billets or bars and is accompanied by a large increase in its length and by an equivalent reduction of the other two dimensions. On the other hand, when rolling into slabs, plates or sheets or in the forging of flat pieces or plates, the same constituent elements of the steel undergo deformations which still consist mainly in “stretching” in a longitudinal direction, resulting in a considerable increase of the corresponding dimension, but is also accompanied by a smaller increase in the transverse dimension.

Thus, when rolled into bars, the structural elements assume the general appearance of fibers all arranged with their length in the direction along which the elongation produced by hot-work has taken place. Structural elements in plates however possess the general “likeness” of elongated flakes all arranged with their greater axis parallel to the “main” or “longitudinal” direction and with their width parallel to the transverse direction of rolling or forging.

From the viewpoint of morphology of the structural elements and their colonies, a main direction must be distinguished in a rolled rod parallel to the elongation produced by hot-work, and is called “longitudinal” to differentiate it from other directions normal thereto, any of which may be called “transverse.” In the case of plates, however, two *principal* directions must be distinguished, the first of which is parallel to the direction of maximum elongation and is called the “main longitudinal,” while the second is parallel to the widening deformation and is called the “main transverse” direction.

Such morphological anisotropy of structural elements in rolled or forged steels produces an anisotropy of mechanical properties, and because of this, various mechanical tests imposing stresses directed along the *main* direction differ remarkably from those measured transversely. This is obvious when we think of the close connections shown

nents in them, and such like characteristics. In hot-worked steel we are confronted with the same facts taken under consideration in former chapters when studying the effect of heat treatment on ingotism, characteristic of castings.

In a very rudimentary and purely intuitive way, we may think that the longitudinal stresses have a tendency to displace the structural elements of the steel one from the other, causing intracrystalline and intercrystalline slip in a direction parallel to that of their best-developed surfaces of contact; while transverse stresses have a tendency to separate said structural elements *without slipping*. In the latter instance, the simple breaking of contact in a small portion of one of those adhering surfaces causes a fissure *normal to the direction of the stress*. That is to say, it is so placed as to be apt to become a "breakage primer" splitting open the elementary surface on a continuance of the stress, and causing a rapid and very large decrease in the adhesion between the two structural elements considered.

I would be digressing a great deal from the principal theme of my study if I were to pause at this place and show how the differences between longitudinal and transverse mechanical properties of rolled and forged steels may be very plausibly explained by the theory of intercrystalline amorphous cement proposed by Beilby, and accepted by many metallographists. Neither is this the right place to develop an analysis, no matter how interesting or important, of the distinguishing characters of the two groups of properties. Usually these are briefly indicated with the two expressions, which shall be used hereafter, of "*longitudinal mechanical properties*" and "*transverse mechanical properties*." Besides, the examples which shall be mentioned later on, will be sufficient to give a quite precise and complete idea of the differences between the two.

It is necessary to remember that the mechanical properties of rolled slabs, plates and sheets or flat forged pieces or plates determined in the main transverse direction are numerically intermediate between those in the longitudinal and in the direction of squeezing, that is, the transverse direction, properly speaking.

The ensemble of poor transverse mechanical properties—especially in a steel in which these differ considerably from the longitudinal properties—constitutes a net result of the greatest practical importance. In shop parlance "transverse weakness" (Italian "transverso") is usually given to this character of forged or rolled steel. In a general way it may be stated that a transversely-weak steel possesses transverse mechanical properties which are always a great deal *inferior* to the longitudinal

istic "woody" fracture of the tension test-pieces, a typical example of which is reproduced in Fig. 160.

As we shall clearly see later on, the same homogeneity heat treatments which are utilized in cast steels to eliminate ingotism may be employed to decrease or eliminate transverse weakness in forged or rolled steels. It appears that the ultimate crystallographic causes of ingotism and transverse weaknesses are the same.

171. Inasmuch as the previous chapters have shown that the arrange-



FIG. 160.—Typical woody fracture.

ment and orientation of neighboring colonies of secondary crystalline elements resulting from the  $\gamma \rightarrow \alpha$  transformation are strictly bound to the mutual arrangement and orientation of the primary crystalline elements, it is easy to expect that the phenomena which we have briefly examined in the previous section must produce analogous phenomena of congruent orientation in the secondary crystallites of  $\beta$  and  $\alpha$  ferrite and pearlite.

Experience fully confirms these obvious expectations. Hence, it follows that hot distortion produces effects upon the orientation of neighboring secondary crystalline elements similar to those observed when working has ceased before the piece has reached  $A_{r3}$ , not only in consequence of the phenomena which are collected in group 4 of Sec. 168, but also because of those gathered together in group 2.

In all commercial steels the differences in concentration of the

various elements resulting from the primary solidification of the steel persist (at least in part) in the  $\gamma$  crystallites at the start of the transformation. From the considerations developed in the previous chapters of Parts II and IV in this regard it is therefore easy to understand that the arrangement of the colonies of  $\beta$  and  $\alpha$  iron shall copy the general forms assumed

through the transformation interval, we find crystalline elements of  $\beta$  and  $\alpha$  ferrite gathered in groups like *filaments* or closely piled bars ("columnar" structure). On the other hand the same elements of ferrite are gathered in groups like parallel plates ("lamellar" structure) in case the steel has been rolled into slabs, plates or sheets or forged in flat pieces or plates. Thus, Fig. 161 shows at 100 dia. the lamellar structure of rolled mild steel.

**172.** It is obvious that the phenomena which have been considered in the previous section will manifest greater intensity the stronger the heterogeneities persisting in the  $\gamma$  mixed crystals immediately above  $A_{r3}$ , and the slower the cooling through the transformation interval.

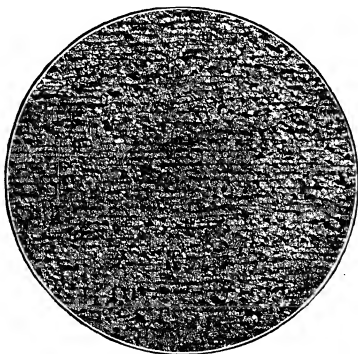


FIG. 161.—Lamellar structure in rolled mild steel.  $\times 100$ .

Likewise obvious is the explanation that the formation of columnar ferrite and lamellar ferrite may also be prevented—or at least, slowed down and rendered less complete—by means of the same homogeneity annealings and quenchings which are utilized to limit or prevent the formation of the wide network of ferrite.

Figure 162 shows at an enlargement of 100 dia. the microstructure of a rolled soft carbon steel, in which the lamellar structure, at first highly developed, has been suppressed by means of a homogeneity heat treatment.

The rationale of heat treating hot-worked pieces is plain, keeping in sight what has been attempted to be shown in the previous sections:



its influence upon the mechanical properties of rolled or forged steel to that of the deformation of the primary crystalline elements, in a manner analogous to cast steels, where the influence of the larger network of ferrite accrued to that due directly to the primary crystalline elements. Therefore, as in the case of cast steels where a highly developed wide network of ferrite aggravated the phenomena of ingotism, a high development of a columnar system of ferrite in forged or rolled steels causes a corresponding deterioration in quality due to the phenomena of transverse weakness.



FIG. 162.—Partial suppression of lamellar structure in Fig. 161 by heat treatment.  $\times 100$ .

Consequently, one may foresee that the application of homogeneity heat treatments to rolled or forged steels may be regulated by criteria entirely analogous to those regulating the normalization of cast steels. While for these latter the main result consisted in the elimination of ingotism, the preliminary heat treatment of rolled or forged steels mainly eliminates the troubles inherited from the sum of all those characters which we have indicated with the name of transverse weakness.

The examples which will be given later will prove the correctness of these expectations.

Meanwhile, it may be pointed out that the foregoing remarks will permit me to avoid dwelling upon arguments purporting to show how the interpretation of reported data may be clearly based upon what has been

said in the first two parts of this volume, inasmuch as these arguments would only be a repetition of those developed in Part IV, and which may be completely applied to the new cases under discussion with the only exception that account must be taken of the observations made immediately above.

**173.** We can clearly predict that the presence of emulsified sonims in forged or rolled steel, especially when in ranks and strongly oxidized, facilitates the development of high transverse weakness and of columnar and lamellar ferrite, and stubbornly counteracts the effects of homogeneity heat treatment.

This is evidently due to the same reasons by which we ascribed to the presence of the same inclusions in cast steels the more tenacious persistency of the phenomena of ingotism, the thick shells of ferrite, and the strong tendency for the same network to re-form during cooling after an annealing.<sup>1</sup>

As shall clearly be shown by examples, this prediction is also fully confirmed by experience. Therefore it will not be necessary to repeat later any special discussion of the causes of persistent transverse weakness due to sonims, nor of the special development of lamellar or columnar structure due to the same inclusions.

The practical importance of the effects produced in rolled or forged steels by oxidized emulsified inclusions is greater than in cast steels. This is due mainly to the fact that in the former the effects directly or indirectly produced by the non-metallic inclusions (especially upon the development and persistency of transverse weakness), are added to those produced directly by the phenomena accompanying hot-work, belonging to the four groups indicated in Sec. 168, often accentuating their consequences very strongly.

We shall also see how it sometimes happens that the just mentioned characteristic effects of oxidized sonims appear with great intensity in forged or rolled steels, yet are caused by inclusions so small in dimension as to be difficultly detected under the microscope, especially if more or less deeply etched surfaces are examined. In some of these cases it is difficult to certify the exact cause for defective metal. Doubts in this regard are often resolved by microscopical examination under high magnification of a perfectly polished, unetched section. Under these conditions the field often reveals extraordinarily numerous inclusions, but of exceedingly small dimensions.

**174.** Phenomena of the third group mentioned in Sec. 168 have already been discussed briefly in Parts I, III and IV (see for instance Sec. 82, 93, 94). It is not appropriate to study peculiarities of the phenomena in question here because, as has been pointed out many times, their effects are not felt in a considerable measure in modifying diffusion phenomena

although having a great practical importance upon the mechanical properties of steel. As a rule, therefore, crystalline slip and twinning do not intervene in noteworthy measure in studying modifications in heat-treatment made necessary by previous hot-work. This detailed crystallographic study should form the subject matter of a most important chapter in any complete treatise on the metallurgy and metallography of steel. It is to be noted with regret that in some important works covering these branches of science, sufficient space has not been given to the study of plastic deformation and its effects, space which is deserved by the great practical importance which such phenomena present in numerous commercial processes such as rolling, forging, pressing, drop forging, hot and cold drawing, etc. However, it is proper to add that this serious gap has been filled in many of the most important and recent publications, as those by H. M. Howe, Albert Sauveur, and G. Tammann.

More than one hypothesis based upon wide and sound experimental foundation have been proposed to explain the very remarkable direct effects produced by the phenomena of this group. We hardly have room here to dwell upon such hypothesis, prominent among which Beilby's ideas about amorphous cement—although subject to serious objections—appear to be founded upon concrete and direct observations.

The experimental data to be mentioned later in this Part will furnish a sufficiently complete and precise idea of the great practical importance of plastic deformation and of the amount of its *direct* effects upon the mechanical properties of the steels with which we are concerned. This will be especially apparent when such data will be compared to those mentioned in Part IV regarding the mechanical properties of cast steels, variously treated. It will then be clearly evident that the *direct* effects of these phenomena ordinarily appear as a remarkable improvement in toughness or resistance to impact. Such comparison will also especially emphasize the fact that the effects of crystallographic slip and twinning act directly upon the properties of the metal and do not practically consist in modifications upon the course of homogeneity heat treatment such as those produced by hot plastic deformations. In fact we shall see that in the great majority of normal rolled or forged steels, even before being subjected to any particularly energetic homogeneity heat treatments, there occurs a great decrease in fragility or in what amounts to the same thing, a great increase in the resistance to impact, when their physical properties are compared to those of the original ingot.

It may be observed immediately that the resiliency is at a maximum when it is measured by an impact test taken normal to the elongation produced by forging or rolling. That is to say, a longitudinal test will give much higher values than a transverse test. If the impact flexure

duced by hot-work and, therefore, tested with an impact directed in a direction parallel to that of said elongation, the results will be low.

Concerning other mechanical properties, including resistance to tensile impact, we shall see by comparison with the data referred to in Part IV that the improvements produced by the phenomena of the third group mentioned in Sec. 168 are not greater than those which can be obtained with cast steels of properly selected composition, after having been subjected to a rational homogeneity annealing, occupying no greater time than that required for rolling or forging. In other words: as far as those other mechanical properties are concerned, a steel casting subjected to a rational homogeneity heat treatment will possess characteristics entirely due to the phenomena of slip and twinning we are now discussing equal to those which it is possible to obtain with forged or rolled steels.

175. Before passing on to the examination of the phenomena pertaining to the fourth and last group specified in Sec. 168, it is necessary to note some more precise information regarding the processes which take place in those cases, already mentioned in the same Section, in which the crystallographic phenomena continue to develop, even after mechanical work has ceased exercising a new specific action. Such crystalline growth occurs during the heating processes constituting the preliminary heat treatment. As has been already pointed out, such cases could be erroneously explained by saying that the continuing crystallographic modifications are also due to the phenomena of diffusion responsible for a veritable normalizing. In reality, grain growth is still a direct effect of the plastic deformation occurring during the previous mechanical treatment. On account of this possible misinterpretation it is now desirable to present some experimental data illustrating the specific characters of such crystalline changes. Thus we shall avoid repeating a discussion in later chapters to clear up the apparent anomalies in the heat treatment of rolled steels due to the appearance of such phenomena.

The fundamental result is a large increase of the dimensions of the primary crystalline elements of the steel or of their colonies, *i.e.*, a large "grain growth." This increase occurs when a steel is heated after it has undergone plastic deformation responsible for intracrystalline twinning and slip along the planes of cleavage, which occurrences have been gathered into the third group specified in Sec. 168. Within certain limits of temperature and deformation so well determined for each steel as to constitute actual *critical intervals*, the increase in grain size is greater the more intense has been the crystallographic rearrangement, the higher the temperature of subsequent heating, and the longer this heating has been protracted.

Inasmuch as the amount of slipping and twinning is greater, other conditions being equal, the lower the temperature at which plastic deformation of a definite amount has been accomplished, whenever we

consider only the above mentioned critical intervals it can be predicted that grain growth will be of greater importance when the hot-work has been continued to subnormal temperatures.

Experience fully confirms this prediction, and shows that grain growth is the most notable, other conditions being equal, when plastic deformation has taken place at ordinary atmospheric temperatures, as is done for instance in cold stamping, cold rolling or cold drawing. Furthermore, in the ordinary hot-working processes grain growth is greater the colder the piece has been finished.

It is well known that if a steel undergoes plastic deformation at relatively low temperatures, temperatures whose upper limits vary widely according to the composition of the steels but which usually do not extend beyond incipient red, it simultaneously acquires quite different physical properties which are ordinarily thought of as being peculiar to "cold-worked metal." Normally it possesses a higher elastic limit and tensile strength than before.

When the stresses which have caused plastic deformations have taken place either below or above the just mentioned critical temperature interval, or when an annealing temperature has reached values below or above the extreme critical points, the phenomenon of grain growth becomes less intense.

**176.** From the viewpoint of purely structural characters the effects produced by grain growth upon annealing cold-worked metal make themselves evident only in very soft steels containing between 0.05 per cent. and 0.15 per cent. carbon. All the facts indicated in Sec. 175 are exhibited in the single plane section illustrated in Fig. 163, reproduced from a photograph published by Chappell. This represents a cylindrical test-piece of very soft steel (0.06 per cent. carbon) broken in tension and then annealed for about 2 hours at  $600^{\circ}\text{C}.$ , then split down the axis, polished and etched.

In fact, it is possible to determine by examination of this sample the effects produced by different specific loads by noticing the tapered form assumed by the test-piece under stress and knowing the total value of said stress.

Thus we see at the bottom of the figure that grain growth has been practically zero for the sections larger than a certain limit, namely, for unit loads below a certain value. Immediately beyond this limit, the grain growth becomes the most pronounced, whereupon it decreases gradually step by step as the section decreases (and therefore step by step as the unit load increases) until again becomes practically zero for sections near the breaking point of the sample.

Figure 163 therefore exhibits the whole critical interval of plastic

The practical importance of these critical intervals as regards the selection of heat treatments for cold-worked steels is quite evident.

Speaking from the standpoint of the mechanical properties of the steel, it is sufficient to notice that grain growth ultimately may be considered as a concordant re-orientation of the crystalline elements constituting several contiguous grains, so that the new arrangement of the planes of cleavage of such adjusted grains brings them parallel with the prolongation of those of several adjoining grains. It is easy to foresee that such an enlargement of the planes of cleavage, an enlargement which may reach very great values, must give rise to noteworthy modifications in the mechanical properties.

Again this last remark is fully confirmed by experience. It has also been demonstrated that the characteristic modifications in the mechanical properties of steel due to heating after cold-work may occur with great intensity even for those high carbon steels in which an increase of the grain is not revealed upon microscopic examination, or if visible, but imperfectly.

For oft repeated reasons it is impossible to pause and examine in detail the extremely various manifestations produced by the modifications under discussion. I will therefore confine myself to mentioning an example which shows the effects of grain growth with particular clearness.

Figure 164 illustrates at about half the natural size the characteristic appearance of the long narrow fissures produced in a field-gun shield after being hit by rifle bullets fired against it in a direction normal to its faces. The thickness of the plate was 4 mm. and the bullet was the usual Italian service ammunition, clad in a nickel-copper sheath. The shield was hit at a velocity of about 670 meters per second.



FIG. 163.—Grain-growth in annealed tension specimen. (Chappell.)

Many plates were obtained from ingots of the same heat by a similar and closely observed process of rolling, followed by homogeneity quenchings, and then by final quenchings and drawings. When proof-tested they showed that the characteristic impact fissures of Fig. 164 only occur in plates whose rolling had proceeded at lower temperatures. The great

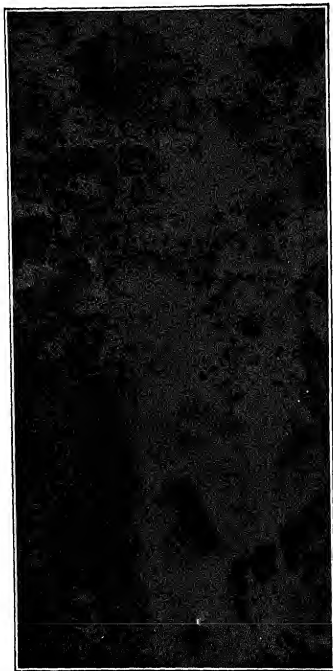


FIG. 164.—Gun shield cracked by bullet impact. Half size.

development of planes of cleavage clearly appears in this instance to be due to the heatings during the various heat treatments after the cold-work

instruments, or aeroplane construction, after the first cold drawing the wire is subjected to homogeneity heat treatments often alternated with further cold drawings. Excessive grain growth may thus occur often accompanied by very serious deterioration in the mechanical properties of the wire, primarily due to the fact that the deformation suffered by the metal during drawing has taken place just within the critical intervals described in Sec. 175.

Sometimes it happens that for the particular requirements of the manufacturing process it is not possible to avoid or sufficiently suppress the deleterious consequences, except by executing all the homogeneity heat treatments in the best manner possible, *before* the steel has been subjected to any mechanical strains apt to cause the dangerous cold work, renouncing afterwards any further heat treatment. For instance, in the case of high tension wire this normalization, and other thermal manipulation would be done upon the hot-rolled rod.

At any rate, it should be clear ere this how extraordinarily important it is to study accurately the critical intervals of cold work, in relation to the consequences which may be produced during heat treatment. Most especially important it is whenever there is reason to believe that cold straining would follow a given series of hot workings, in turn to be followed by a heat treatment.

I have already briefly mentioned in Sec. 107 the fact that grain growth entirely analogous in microscopic appearance and practical consequences to those which we have studied, may even take place due to the so-called "internal stresses" carried by some parts of the steel pieces subjected to simple heat treatments. These stresses are due to the different variations in volume produced by heating and quenching in the different parts of the same object. As has been pointed out in Sec. 107, the phenomena of grain growth consequent upon internal stresses may, in a certain sense and in these cases, be considered as true effects of heat treatment.

**178.** Finally, we must also briefly investigate the general characters of the phenomena gathered together under group four as defined in Sec. 168. These take place when the hot work partially or wholly occurs within the  $\gamma \rightarrow \alpha$  transformation interval.

As has already been indicated in the same section these phenomena may be compared to the so-called disturbed crystallization, so well known in salt solutions; phenomena which have as essential consequences the entire or almost complete elimination of any under-cooling and a considerable decrease in dimension of the resultant individual crystalline elements.

Recalling the remarks made in Part IV, especially in Sec. 116 and Sec. 139 *e.s.*, regarding the forms and characteristic arrangements as-



sumed by the crystalline elements of  $\alpha$  and  $\beta$  ferrite after undisturbed separation from austenite, I believe that the characteristics of the various phenomena which may appear during disturbed crystallization can be summed up in the following groups and sub-groups:

(4) After its complete solidification the steel has been subjected to a homogeneity heat treatment similar, for instance, to that indicated in Sec. 114, designed to eliminate in the most *complete* way the differences in carbon concentration between the nucleus and the periphery of each primary mixed crystal.

This case happens only very rarely in practice, as has been pointed out several times. In it the process of disturbed crystallization may result in different structures according to the composition of the steel and the velocity of cooling through the transformation interval:

*First.*—If the cooling takes place very slowly and the steel contains less than 0.35 per cent. of carbon, the ferrite crystallization occurs in two periods of time, as we have seen in Sec. 140 (2,a). However, in the case of undisturbed crystallization as there discussed, the lamellæ of ferrite forming *directly* in the  $\alpha$  state appeared as prolongations upon the globular masses of first-segregated  $\beta$  ferrite acting as germs of crystallization. On the other hand, in this present case of disturbed crystallization the nuclear orienting action exercised by the first massive crystals cannot be felt in substantial measure, and the two types of ferrite crystals ( $\beta$  globules and  $\alpha$  lamellæ) form separately from each other. Due furthermore to the known effects of disturbed crystallization, the dimensions of each crystalline element of ferrite appear much smaller, other conditions being equal, than the size of the same elements formed in a similar process of undisturbed crystallization. Finally, the deformations undergone by the entire metallic mass are accompanied by deformations of the crystalline ferrite, which often deeply change its appearance.

The above mentioned limit of 0.35 per cent. of carbon refers to the common carbon steels, as must be clear from all the considerations developed in the previous chapters. For special ternary and quaternary steels, this carbon limit is lowered in accordance with the criteria and data presented in the Part I and II; the indicated phenomena vanish for the special steels called "high alloy," in which the separation of the two forms of ferrite is no longer distinct.

*Second.*—If the cooling through the transformation interval takes place very slowly and the steel contains more than 0.35 per cent. carbon (or more than the corresponding amount when it is a special steel) the ferrite will be entirely of the lamellar type; but the laminations will be a great deal smaller than in the case of undisturbed crystallization. Furthermore, simultaneous deformation with ferrite cryalli is reflected

*Third.*—If the cooling through the  $\gamma \rightarrow \alpha$  transformation takes place very rapidly, as for instance, in drop forging small pieces, the various crystalline elements appear in very small dimension, and imperfectly developed as to form, so that it is difficult to distinguish one transformation product from the others.

(B) The solidified steel has been subjected to a heat treatment insufficient to totally obliterate the differences in carbon concentration between the nucleus and the periphery of the individual primary mixed crystals.

We have already seen that this case is far more common. In it the course of disturbed ferrite crystallization may take place in the following



Fig. 165.—Alpha and Beta ferrite in 0.23 per cent. carbon forging.  $\times 100$ .

different ways, according to the cooling velocity and the composition of the steel:

*First.*—If the cooling is very slow and the steel contains less than 0.35 per cent. carbon (or less than the corresponding amount in a special steel) the structure of the completely cooled metal is still similar to that which we have described for the case A, above, except that in the present case traces of the large network of ferrite may remain, more or less deformed. The more or less marked tendency for the large network of ferrite to persist permanently depends upon the same causes which have already been examined in Part IV.

Figure 165 reproduces at an enlargement of 100 dia. the structure of such a 0.23 per cent. carbon steel, in which no appreciable traces of the

large network of ferrite have remained. The figure shows clearly the two types of ferrite crystals independent of each other.

*Second.*—If the cooling through the  $\gamma \rightarrow \alpha$  transformation interval is still slow, but the steel contains considerably more than 0.35 per cent. of carbon (or more than the corresponding value for a special steel), the structure of the cold metal will still be like that described for A, 2, above namely: the elements of lamellar ferrite will still predominate, more or less altered in form by the deformations suffered by the metallic piece. Contrary to what happened in the case A, 2, there may also appear a greater or less quantity of globular ferrite formed in those regions where the local carbon concentration was actually less than 0.35 per cent. due to persisting heterogeneities in various parts of the primary mixed crystals.



FIG. 166.—Steel of Fig. 165 after forging down through transformation interval.  $\times 100$ .

*Third.*—If the transformation interval is crossed rapidly during mechanical work the resulting structure is practically identical to that formed in the case A, 3. An example of this case is given by Fig. 166, which reproduces at 100 dia. the microstructure of the same steel as Fig. 165, after being cut into bars about 40 mm. dia. and forged down until reduced to 35 mm. dia.

All the observations contained in this section apply only to those cases in which the mechanical work is executed in a normal and correct manner so as to cause a *gradual* deformation of the steel (*i.e.* forging or

greatly reduce the initial dimensions of the ingot or billet, the resulting structures almost always belong to the types characterized by either columnar or lamellar ferrite.

Even during the development of the phenomena of the fourth group of Sec. 168 the phenomena of the second group continue, and for the same reasons which we have already studied. The effects of such phenomena of "orientation" in the primary crystalline elements and their colonies add to the effects of forging.

## CHAPTER XX

### EFFECTS OF PRELIMINARY HEAT TREATMENT UPON FORGED OR ROLLED STEELS

179. Having thus briefly examined the general characters of the phenomena through which the effects of hot-work make themselves directly or indirectly felt upon the course of homogeneity heat treatments and upon the results which can thereby be obtained, we may now present data concerning some of the cases which may ordinarily occur when heat treating rolled or forged steels.

We shall in no case need to dwell at length upon the interpretation of the facts which shall be mentioned, either in an explanation of their causes or practical importance, because the conclusion would always be based on the criteria developed in the previous chapter. For instance, we will find many examples among the experimental data which I shall have occasion to relate, which confirm the fact that the heterogeneity due to ferrite segregation during the  $\gamma \rightarrow \alpha$  transformation may exercise an equal and even greater influence upon the mechanical properties of steel than that of the chemical heterogeneities in the primary mixed crystals due to the process of solidification. This remark applies especially to the fact that forged or rolled steels are much more susceptible to the influence of homogeneity *annealings* than of homogeneity *quenchings*. Besides, the same considerations may be adapted to forged or rolled steels as have been noted when discussing cast steels, as far as concerns the practical application of the criteria upon which are based the processes of *attenuated quenching* (or air hardening). Likewise we shall be able to apply to the new cases the old observations regarding the practical utilization of thermal hysteresis.

The examples shall be selected so as to illustrate the preceding general discussion in as orderly and complete a manner as possible under the circumstances of limited space.

180. In order immediately to place in the best possible light the differences and analogies between the properties of cast normalized steels and the same steels after being forged and then given the same heat treatments, it is opportune to mention some examples in which those differences appear with great intensity. It is clear that those steels must be exhibited which experience has demonstrated to be less

To this group belong many nickel-chromium steels. As a first example I select one of the following composition:

Carbon.....	0.28 per cent.
Manganese.....	0.32 per cent.
Silicon.....	0.16 per cent.
Phosphorus.....	0.03 per cent.
Sulphur.....	0.02 per cent.
Chromium.....	0.85 per cent.
Nickel....	2.53 per cent.

The following table contains notes on the treatments to which the steel has been subjected and the corresponding mechanical properties.

No.	Mechanical work	Heat treatment	Static tension test				Impact resistance (Charpy test), kg.-m. per sq. cm.	Static bending	Micro-structure
			Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.			
1	none	Annealed for 10 hours at 900°C., followed by slow cooling in the furnace; afterwards drawn at 680°C. for 6 hours.	86,300	49,800	12.0	11	2.52	test-piece broken	Fig. 167
2	forged at 900°C. to 1100°C.	Annealed and drawn under identical conditions as above.	87,500	56,900	27.3	56	9.9	not broken	Fig. 168
3	forged at 900°C. to 1100°C.	Annealed at 900°C. for 10 hours followed by a quenching in water at 20°C.; afterwards drawn at 680°C. for 6 hours.	98,000	72,500	23.0	62	15.6	not broken	Fig. 169

Tests 2 and 3 were made on longitudinal tension and impact test-pieces, *i.e.*, with their axis parallel to the elongation produced under the hammer. The forged blocks were 95 × 95 mm. in cross section and were reduced from raw ingots 200 × 200 mm. in section. Forging was done under a 1000 kg. hammer. The various physical tests were made upon pieces having the forms and dimensions indicated in Sec. 151.

Hot work is thus shown by the foregoing table to influence the

described in Sec. 152<sup>1</sup> where the experiments were made upon steels better adapted for castings than for forged or rolled pieces.

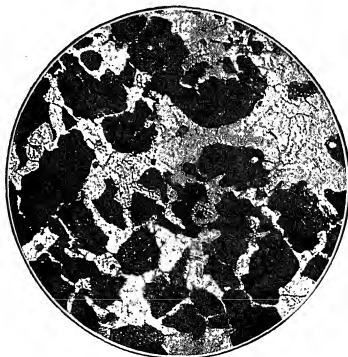


FIG. 167.—Heat-treated nickel-chromium steel casting.  $\times 100$ .



FIG. 168.—Same steel and same heat treatment as Fig. 167, except that piece was forged.  $\times 100$ .

Comparison of the second treatment with the third furnishes an example of the characteristic differences between the effects produced

by homogeneity annealings and those produced by homogeneity quenchings upon the mechanical properties of a forged steel. In regard to this

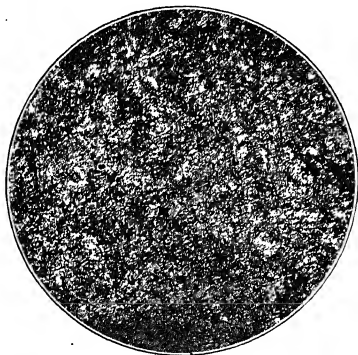


FIG. 169.—Structure of nickel-chromium forging after normalizing, quenching and drawing.  $\times 100$ .

second point, it is especially necessary to remember that this present comparison is between physical properties of *longitudinal* test-pieces,



FIG. 170.—Transverse fracture of forged nickel-chromium steel.

because the differences would be a great deal greater if the test d



proves that the differences in mechanical properties are really due to the phenomena which we have summarized in the previous paragraphs of this chapter. Without repeating things already said, a comparison of Fig. 168 with Fig. 169 confirms the remark of Sec. 171 that slow cooling of a forged or rolled steel following a homogeneity heating may easily re-form a columnar or lamellar ferrite system after the chemical heterogeneity existing in the primary mixed crystals, has been drawn out by the previous hot work. A structure such as appears clearly in the second treatment has not been able to develop in the third because it has been suppressed by quenching.

Figure 170 illustrates the fracture of a cylindrical tension test, taken *transversely* from the steel treated as noted in No. 2 of the above table. Characteristic signs of transverse weakness can still be seen clearly upon this surface.

Here we have a first example of the characteristic differences between the effects which homogeneity annealing may produce upon the mechanical and structural properties of forged steels as compared with those which homogeneity quenchings produce upon the same steels, even in the cases in which the two treatments include practically identical heating periods and, therefore, almost equal chances to equalize the chemical concentrations by diffusion.

We shall presently see other examples of the same fact.

181. In those compositions which are especially adapted for heat treated steel castings<sup>1</sup> the *direct* effects produced by hotwork, especially upon impact strength, are not only smaller numerically than those exhibited in the last table but the differences between the effects produced by homogeneity quenchings are also smaller in comparison with those which may be obtained by corresponding annealings. This is obvious in view of the familiar fact that the effects of annealing or quenching executed after equivalent conditions of heating differ mainly because of the different course of ferrite segregation, a course which in turn depends essentially upon the frequency of crystallization germs appearing in the  $\gamma$  solid solution.

As an example of this last fact let us take the case of a nickel steel of the following composition:

Carbon.....	0.34 per cent.
Manganese.....	0.57 per cent.
Silicon.....	0.25 per cent.
Sulphur.....	0.02 per cent.
Phosphorus.....	0.03 per cent.
Nickel.....	2.00 per cent.

Experiments upon such steel gave the results in the following table:

and refer to tests made under identical conditions with those given in the previous section.

Mechanical work	Heat treatment	Static tensile test				Impact resistance (Charpy test), kg.-m. per sq. cm.	Static bending	Micro-structure
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.			
none	Reheated for 10 hours at 900°C. followed by slow cooling in furnace and then drawn at 680°C. for 6 hours.	78,800	44,100	22.5	40.5	7.25	Test-piece remained intact	Fig. 171
none	Reheated for 10 hours at 900°C., followed by slow cooling in furnace, then by an annealing at 780°C. for 6 hours. Finally a slow cooling in furnace.	78,100	44,100	20.5	30.0	4.1	Test-piece broke	Fig. 172
forged at 900°C. to 1100°C.	Reheated at 900°C. and drawn at 680°C. under identical conditions to those of test No. 1.	80,000	48,400	23.0	51.0	9.5	Test-piece remained intact	Fig. 173
forged at 900°C. to 1100°C.	Heated at 900°C. during 10 hours followed by quenching in water at 20°C. and then by a drawing at 680°C. for 6 hours.	84,600	56,000	24.0	54.0	12.3	Test-piece remained intact	Fig. 174

Examination of the data contained in the table and the microstructures reproduced in Fig. 171 to 174, at an enlargement of 130 dia., fully confirms what was said above.

To the three heat treatments noted in the previous section I have added No. 2 of the present table which ends with a true annealing. One can now compare its results with No. 1, which ends with a simple drawing. It is clear that a simple draw suppresses the second segregation which takes place in No. 2 during the slow cooling following the second reheating. Thus the effects of the second segregation show

the slight diffusion which takes place—at least in regions richer in carbon—while drawing at  $680^{\circ}\text{C}$ ., a diffusion whose effects cannot be further modified during the following slow cooling.



FIG. 171.—Nickel steel casting, after heat treatment ending in a simple draw.  $\times 130$ .

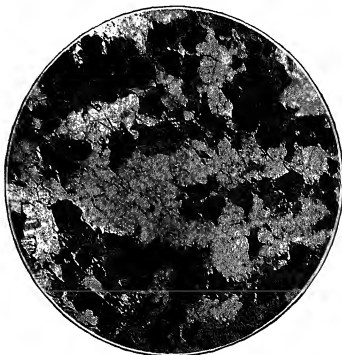


FIG. 172.—Nickel steel casting, after heat treatment ending in a true anneal.  $\times 131$ .

These last conclusions are completely confirmed by comparing the structure obtained by means of the first treatment, Fig. 171, with that produced by the second treatment, Fig. 172, clearly showing the segregation of ferrite.

All the other data contained in the table require no comment, since they are sufficiently clarified by previous discussions. It may be added that ferrite segregation is evidently the cause of the differences in the results of treatments No. 3 and 4.



FIG. 173.—Nickel steel forging, annealed and drawn.  $\times 130$ .

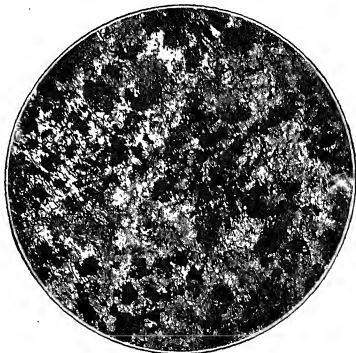


FIG. 174.—Nickel steel forging, annealed, quenched and drawn.  $\times 130$ .

182. In order to complete this first group of comparisons between the effects of identical mechanical and heat treatments, a series of experiments were conducted on a series of specimens of nickel steel forging, which were subjected to identical mechanical and heat treatments, and then examined microscopically.

in the two previous sections upon the physical properties of a very soft carbon steel. As has already been pointed out, the structure of steels of this type is strikingly modified by the phenomena studied in the previous chapter.

The steel in question had the following composition.

Carbon.....	0.08 per cent.
Manganese.....	0.79 per cent.
Silicon.....	0.10 per cent.
Sulphur.....	0.02 per cent.
Phosphorus.....	0.04 per cent.

The following table contains data covering the treatments to which the steel was subjected and the resulting physical properties. As remarked above, the forging, heat treatment and mechanical tests were

No.	Mechanical work	Heat treatment	Static tensile test				Impact resistance (Charpy test), kg.-m. per sq. cm.	Static bending	Micro-structure
			Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.			
1	none	Reheated for 10 hours at 900°C. followed by slow cooling in furnace and then drawn at 680°C. for 6 hours.	55,200	34,100	29.5	56	7.6	Unbroken	Fig. 175
2	none	Reheated for 10 hours at 900°C. followed by slow cooling in furnace, then by an annealing at 780°C. for 6 hours. Finally a slow cooling in furnace.	54,200	35,600	31.0	59	7.0	Unbroken	Fig. 176
3	forged at 900°C. to 1100°C.	Reheated at 900°C. and drawn at 680°C. under identical conditions to those of test No. 1.	56,300	37,000	31.0	75	19.9	Unbroken	Fig. 177
4	forged at 900°C. to 1100°C.	Heated at 900°C. during 10 hours followed by quenching in water at 20°C. and then by a drawing at 680°C. for 6 hours.	57,200	38,400	28.5	74	20.0	Unbroken	Fig. 178

made under absolutely identical conditions with those mentioned in the two previous sections.

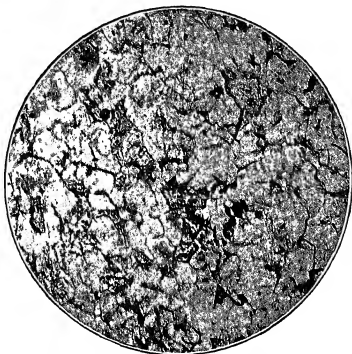


FIG. 175.—Low carbon steel casting after heat treatment ending in a draw.  $\times 100$ .

Comparison of the tabular data and the corresponding microstructures —reproduced in Fig. 175 to 178 at an enlargement of 100 dia.—furnish

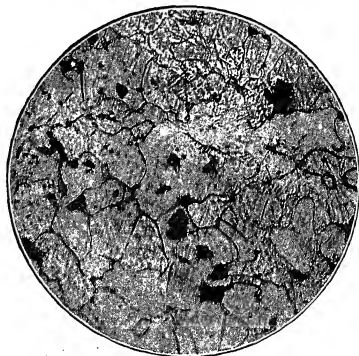


FIG. 176.—Large grained ferrite in steel like Fig. 175 after annealing.  $\times 100$ .

interesting illustrations paralleling those of the previous sections, espe-

In particular, and without repeating in detail what has already been said, the previously noted fact is confirmed that physical properties



FIG. 177.—Traces of cleavage plains intersecting polished surface parallel to extension forging.  $\times 100$ .



FIG. 178.—Forged low carbon steel after complex heat treatment.  $\times 100$ .

We have already pointed out Sec. 135 that this circumstance permits safe use of common steel castings even as substitutes for forged or stamped pieces whenever the requisite elastic limits or other mechanical properties are moderate enough to allow the use of very soft steels. Whenever very high tensile strengths and elastic limits are required, however, such substitution can be made only by using special steels subjected to accurately controlled heat treatments.

One observes upon comparison of the microstructure after treatment No. 1 (Fig. 175) and that of the same steel after No. 2 (Fig. 176) that the former discussion of the two corresponding samples shown in Fig. 171 and 172 is confirmed. In particular large grained ferrite is quite notable whenever the heat treatment ends with a true annealing.

Finally, Fig. 177 shows clearly some light striations in the grains of ferrite revealed by the nitric acid etching. Such lines correspond to intersections of planes of cleavage with the plane of polish, and are now closely parallel, yet they do not have concordant directions grain to grain in the unforged steel shown in Fig. 176. In Fig. 178 this evidence does not reappear because the microsection is normal to the direction of elongation produced by forging, while the previous photograph was taken parallel to that direction.

We have already analyzed the causes of all these phenomena in previous sections.

**183.** The effects produced by homogeneity annealings and quenchings may thus differ greatly according to the velocity of cooling which terminates the heating, and this always in the sense that a more rapid cooling favors the maintenance of the homogeneity obtained by heating above  $A_{c3}$ . This now familiar fact takes on peculiar importance when heat-treating pieces of very large dimension and massive form.

In such cases, it is not only impossible to cool the pieces with the same velocity in all parts, but in some parts of the metal it is even impossible to reach very high cooling velocities. Such limitations depend essentially upon the high thermal conductivity of steel. It is clear that the maximum cooling velocity at any point within the metallic piece will practically be reached when the external surface of the object is constantly maintained at the temperature of the quenching medium. In commercial operation the most drastic quenches are given by water at atmospheric temperature, and for quickest cooling the process generally requires the use of enormous quantities of liquid.

When treating pieces of very large dimension, the maximum cooling velocity which can practically be reached in the deepest regions is often below the minimum cooling velocity necessary to cause a true quenching. This is rather an indeterminate speed, but if quenching be de-



ture, then we know that this minimum cooling velocity for quenching may vary within quite large limits according to the composition of the steel.

At any rate, it is clear that the effects produced by a homogeneity quenching will be greatest at the surface of any piece, large or small; they will diminish gradually toward deeper strata of underlying metal. Finally, in metal contained in the regions whose distance from the external surface of the piece is above a certain figure, the results of quenching will correspond to those of a true homogeneity annealing.

Experience fully confirms these predictions in all the cases where subsequent heat treatments are missing or consist of such low drawings as will in turn cause no new heterogeneities which might accrue to the first ones. In practice, however, variations in the resulting structure are far from showing the regularity from one point to another which could be expected on the basis of a brief statement of the phenomenon, as such has just been outlined.

Very little precise information is known regarding the causes of the strong heterogeneities produced by heat treatments in various parts of large steel pieces. In general they have been ascribed primarily to irregularities in the conduction of heat from the interior to the exterior part of the metallic mass. Variation in heat transfer would be caused by a lack of uniform dimension and orientation of the structural elements of the steel. This fact in turn depends upon abnormal phenomena due to undercooling which have been discussed when considering primary crystallization.

In the second place it seems certain that the action of internal stresses set up within the piece during mechanical and thermal treatment are also partly responsible for variations in the resulting structure. We have already briefly spoken of the effects of these internal stresses upon heat-treatment.

**184.** I shall now mention some examples showing how the phenomena indicated in the previous sections make their presence felt, and the extent of their influence. These instances will be valuable not only as particular illustrations of the principles noted in Sec. 180 to 182, but also on account of their great practical importance.

A large block of medium carbon steel of the following composition was studied.

Carbon.....	0.42 per cent.
Manganese.....	0.67 per cent.
Silicon.....	0.25 per cent.
Sulphur.....	0.01 per cent.
Phosphorus.....	0.06 per cent.

The piece was of rectangular section measuring 760 × 1400 mm and

reduction in cross-sectional area.

After pressing the block was planed on all sides and then heated during about 20 hours at  $800^{\circ}\text{C}$ . It was then immediately quenched with water at  $16^{\circ}\text{C}$ . by an apparatus drenching the block from all sides with heavy sprays of water, a process lasting nearly 2 hours and requiring about 4000 cubic meters of cold water.

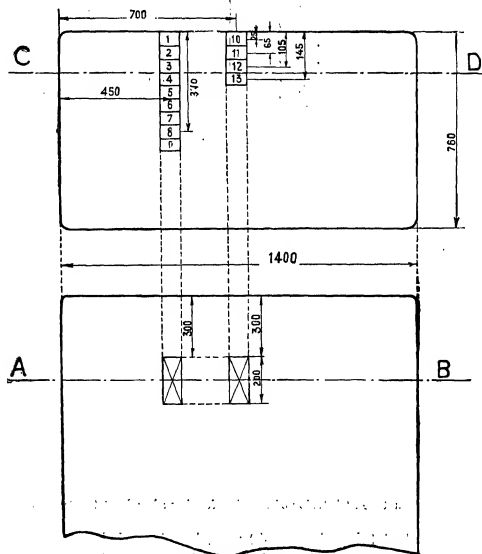


FIG. 179.—Location of test pieces described in Section 184, Section CD below, AB above.

About 20 minutes after the quenching water had been turned off, the temperature of the surface metal had raised to  $100^{\circ}$  to  $120^{\circ}\text{C}$ ., due to transmission of the residual internal heat outward by conduction. Therefore, it was thought superfluous to subject the block to a drawing operation.

Mechanical tests were then made under the identical conditions already described in Sec. 151 upon *longitudinal* test pieces taken from the positions located in Fig. 179; the results of these tests are collected in the following table.

Position No. (see Fig. 179)	Tensile strength, lb. per sq. in.	Elongation, per cent.	Impact resistance (Charpy test), kg.-m. per sq. cm.
3	66,400	16	
4	67,700	10	
5	69,000	16	
6	65,400	17	6.66
7	64,000	18	7.14
8	64,000	18	5.16
9	64,700	17	8.11
10	78,900	24	
11	76,100	22	
12	67,400	17.5	
13	65,000	14	

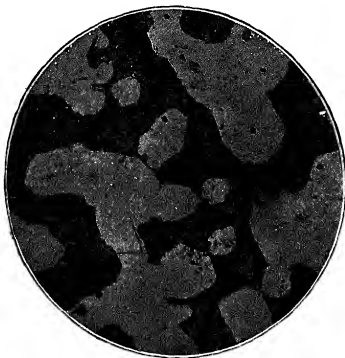


FIG. 180.—Structure at interior of massive forging after heat treatment.  $\times 100$ .

A tension test taken from a 40 mm. bar forged down<sup>1</sup> from a 110  $\times$  110 mm. test ingot taken from the same heat gave the following results:

Tensile strength.....	78,800 lb. per square inch
Elongation.....	21 per cent.

Comparison of this result with those furnished by samples 3 to 13, especially noting that test-pieces 10 and 11 give results closely approximating those from the forged bar, clearly indicates that the effects of pressing the large block and its subsequent heat treatment have been

results, especially in resistance to impact, a phenomenon of which we have spoken in the previous section.

Microscopic examination of the various test-pieces confirms the deduction that the characteristic effects of forging and heat treatment<sup>1</sup> have not been felt beyond a depth of 90 mm. Thus Fig. 180 reproduces at an enlargement of 100 dia. the metallic structure of test No. 12. Such structure is that which a steel of the average composition of the block examined assumes after simply annealing a raw casting, the annealing to be followed by a slow cooling.

**185.** When the forging has been carried further, causing a greater reduction of section of the ingot, its effects are felt to a greater depth, other conditions being equal. As an example, a block substantially identical to the one described in the previous section may be mentioned. This new piece was obtained by pressing an ingot 1200 mm. thick under entirely similar conditions as before, the reduction of the section due to pressing however having been considerably greater than in the previous case, whose original ingot was about 1 meter. The steel was of the same type of the previous block but a little harder grade, containing 0.46 per cent. carbon.

The following table contains the results of a series of physical tests made upon samples cut from the same positions in the new block as shown in Fig. 179.

Position No. (see Fig. 179)	Tensile strength lb. per sq. in.	Elongation, per cent.	Impact resistance (Charpy test), kg.-m. per sq. cm.
3	81,500	18	7.40
4	78,400	15	6.75
5	70,800	16	6.10
6	66,800	10	
7	74,700	10	
8	68,600	11	
9	69,700	10	
10	92,400	23	
11	94,200	21	
12	77,100	20	7.15
13	77,800	15	8.02

It is clearly shown that the combined effects of pressing and heat treatment are more deep seated than in the previous case, noticeably affecting strata as far as 150 mm. from the external surface. Given the fact that the only great difference in the history of the two forgings consists in the amount of pressing which the piece has undergone, it may be concluded that the higher ultimate strength of the metal within 150 mm. of the surface is due principally to the direct action of such hot working.

However, one is not justified in excluding the concomitant influence of an alteration in the rates of diffusion caused by the deformation under the press.

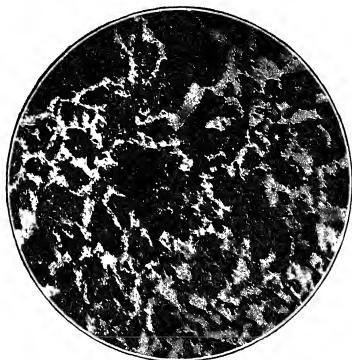


FIG. 181.—Microstructure of test-piece No. 10.  $\times 200$ .



FIG. 182.—Microstructure of test-piece No. 11.  $\times 200$ .

Microstructural examination of the different test-pieces confirms the above precaution. The structures of the three test-pieces No. 10, 11, and 12 are reproduced at an enlargement of 200 diam. in Fig. 181-182.

Taking into account their position in the block, it is clearly seen that the increased agglomeration of ferrite, evidently due to the lesser efficacy of the quenching when passing from the superficial strata to the deeper ones, accompanies the increase in crystalline dimensions. In the light of the history of the block, it is clear that this structural difference is due principally to the different degree of plastic deformation produced by the press upon metal at different distances from the external surface.

**186.** As has already often been pointed out, the phenomena briefly described in the last sections have great practical importance, which is increasing rapidly in step with the growing tendency of the mechanical engineer to require from metallurgical technique metallic pieces of continually larger dimensions, to withstand ever greater stresses. While this



FIG. 183.—Microstructure of test-piece No. 13.  $\times 200$ .

last exigency presupposes the greatest uniformity in mechanical properties in all the parts of the metallic piece, it truly exists only after subjecting the steel to a suitable heat treatment. The importance of studying proper means to eliminate or at least to mitigate the consequences of the phenomena which we have just examined is therefore evident.

When called upon to manufacture pieces of large dimensions which must be subjected to homogeneity heat treatments, best results may be obtained by using steels characterized by a closely packed frequency of crystallization centers, whose transformation range upon cooling lies at very low temperature and consequently which are very sensitive to homogeneous heat treatments. Stable austenite at relatively low temperatures

the rational utilization of thermal hysteresis as mentioned in Sec. 159 assumes special importance since it causes a more or less noteworthy depression of the Ar range.

For reasons already studied, it is possible by such means to prevent or at least largely subdue the ferrite segregation accompanying allotropic transformation, using relatively small cooling velocities in the deepest regions of the piece to be treated. With these precautions the efficiency of the homogeneity treatment is nearly uniform in all regions. The required mechanical properties may afterwards be imparted in an equally



FIG. 184.—Martensite, troostite and massive cementite in one section.  $\times 100$ .

uniform manner by means of a final heat treatment, which may consist in a simple drawing sufficiently prolonged.

As a general example, it may be recalled that it is exactly the rational application of the indicated criteria that permits the fabrication of armor plates of great thickness (sometimes above 400 or 500 mm.) yet equally tough throughout their mass.

**187.** It does not seem to me superfluous to mention here an experiment giving direct proof of the possibility that an occasional steel "takes the quenching" in very different ways in closely adjacent regions. An explanation was noted in Sec. 183 of the non-uniformity sometimes existing in normalized pieces. Yet lacking direct evidence that reasoning might appear unsatisfactory at first sight.

Confirmation is given by the examination of the structure of a high carbon steel (carbon about 1.45 per cent.) which was maintained at 1050°C. about 60 hours then left to cool slowly to about 730°C., whereupon it was quenched in water at 60°C. Figure 184 reproduces the structure of a portion of this steel at 100 dia. It shows a region where the metal retains the mixed structure of martensite and austenite, immediately surrounded by martensite already partially segregated into troostite, in which have even formed various large white crystals of cementite.

Here no doubt can arise that we have discovered two very small contiguous regions in the same mass of steel where the quenching produced very different results, with its corresponding changes in the constitution.

**188.** Finally, before closing this series of observations suggested by the study of the particular effects caused by homogeneity heat treatments when applied to large masses, it is useful to add some experimental data better to indicate methods for practical utilization of thermal hysteresis. As a matter of fact we have just seen a characteristic example of this practical utilization in the case of the preliminary heat treatment of pieces of large dimensions, and I have already indicated the general criteria upon which it is based.

We shall take as an example the heat treatment of a high nickel steel, in which thermal hysteresis is remarkably wide. Besides, this case is particularly interesting from a practical point of view, for the reason that forged pieces manufactured of such steel cannot be quenched from a high temperature without almost inevitably cracking to a greater or less extent and depth. Only the rational utilization of the phenomenon of hysteresis permits one to avoid such damage.

A heat containing 0.33 per cent. carbon, 0.75 per cent. of manganese and 5.80 per cent. nickel was made in the electric furnace, and cast in square ingots  $350 \times 350$  mm.

The tests to be mentioned were made upon round bars about 120 mm. in diameter obtained from these ingots by forging under an eight-ton steam hammer. All test-pieces were *longitudinal* and from the forged bars.

Figure 185 reproduces the differential temperature curves traced with the Le Chatelier-Saladin double galvanometer using a test-piece of the forged steel in question. The bottom curve represents conditions on heating, while the top exhibits the lag on cooling. Without entering in a discussion of the shape of these curves—a discussion which as a matter of fact would be very interesting, especially where it would concern the slow and gradual deviations of the differential galvanometer during the transformation on cooling—it is easy to see that thermal hysteresis is very marked, reaching an amplitude of about 200°C. under the conditions of



Applying the ideas developed in the previous chapters to the numerical data furnished by this curve, it seems evident that the characteristic advantages of a homogeneity quenching will be had in case the temperature has been maintained for a certain time above the transformation temperature (in this case between  $800^{\circ}$  and  $900^{\circ}\text{C}.$ ) and the piece is left to cool slowly and uniformly through an interval of temperature which may amount to almost  $300^{\circ}\text{C}.$ , so that the true quenching operation may then be executed when the entire mass has *uniformly* reached a relatively very low temperature in all its parts. Whenever pieces of this analysis are to be heat treated and are not of exceptionally large dimensions, the quenching temperature may be only slightly superior to  $600^{\circ}\text{C}.$  without

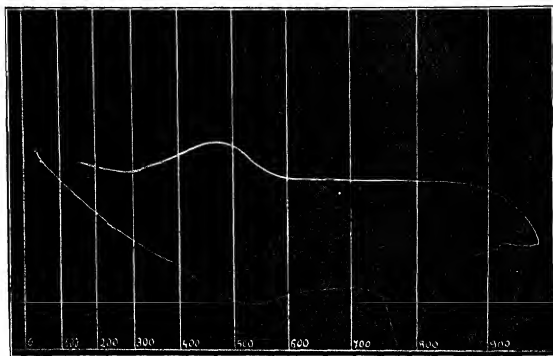


FIG. 185.—Differential curve on heating and cooling a 5.80-per cent. nickel steel.

causing the homogeneity or properties retained by quenching to be noticeably decreased when compared to the physical properties which the same steel assumes after a simple final annealing.

The just indicated fact permits one to utilize fully the advantages of a quenching for homogeneity, even when the form of the pieces to be treated would make it impossible to quench from a high temperature, which would have to be done for a steel having a smaller thermal hysteresis.

In other words: proper utilization of the hysteresis phenomenon permits one to secure the advantages of a homogeneity quenching by means of a heat treatment which approaches more or less a simple homogeneity annealing. This is interesting from the point of view of its applicability to hard steels and pieces of intricate form.

the conditions already indicated in Sec. 151 upon test-pieces obtained in the described manner from 120 mm. forged rounds, quenched and then drawn at 600°C. for 3 hours.

All homogeneity quenching were prepared by heating the bars to 860°C. during 1 hour. Then the bars were left to cool slowly in the furnace until they reached different temperatures indicated in the third column of the table, when they were quenched in oil at 30°C.

No.	Temperature of the pre-heating (during 1 hour), °C.	Temperature of the steel at moment of the quenching in oil, °C.	Temperature of the 3-hour drawing, °C.	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.
1	860	860	600	124,000	107,400	22
2	860	750	600	124,900	107,000	24
3	860	700	600	126,900	108,500	28
4	860	650	600	125,900	107,700	23

The practically complete constancy of results obtained in the four cases proves the foregoing observations to be correct, and demonstrates that a homogeneity quenching will favorably affect the values of the elastic limit to the fullest extent even when the piece is cooled from a temperature only slightly above the minimum indicated by Fig. 185.

## CHAPTER XXI

### COMPARATIVE EFFECT OF ANNEALING AND QUENCHING UPON THE LONGITUDINAL STRENGTH OF HOT WORKED STEELS

189. As a deduction from the considerations which we have so far had occasion to develop and from the specimens which have been described, it can be concluded that the more notable practical advantages of homogeneity quenchings on rolled or forged steels as compared with the "equivalent" homogeneity annealings (*i.e.*, including a heating for an equal period of time and at the same temperature) may be classed in the following two groups:

(a) For steels which have to withstand *longitudinal* stresses a quenching increases the contraction in area more notably than will a slow cooling; the elastic limit and the resistance to impact are also improved.

(b) Steels which have to withstand *transverse* stresses in service have much less tendency to exhibit the phenomena of transverse weakness, and therefore the values for the transverse physical properties more nearly approach the corresponding properties when tested in a longitudinal direction.

In other words: the physical properties—especially the elastic limit, contraction in area and resistance to impact—are improved much more by homogeneity quenchings than by the equivalent preliminary annealings, not only when the piece is tested longitudinally but more especially when the properties are measured *transverse* to the direction of extension.

190. In order to illustrate item (a) of the previous section, I will mention some results of tensile and impact tests made as described in Sec. 151 upon longitudinal test-pieces, cut from round bars about 100 mm. dia. obtained by forging 300 × 300 mm. square ingots.

The data refer to steels of four different heats of similar analysis containing about 0.20 per cent. carbon, 0.50 per cent. manganese, 1 per cent. chromium, and 2.5 per cent. nickel.

The table below collects the results obtained by testing each steel after a simple anneal for about 1 hour at approximately 850°C., and after the same material had been quenched in oil after the same period and temperature in the furnace, and then drawn at about 650°C.

For reference, Fig. 186 reproduces the structure of steel No. 3 as annealed, and the next picture (Fig. 187) the structure of the same steel after quenching and drawing. Both figures are numbered 500. The

ontrasting manner in which the ferrite and the pearlite are segregated gives ample evidence of the quenching efficiency. In accordance with the observation contained in paragraph (a) of the previous section, the rela-



FIG. 186.—Nickel-chromium steel, forged and annealed.  $\times 500$ .



FIG. 187.—Nickel-chromium steel, forged, quenched and drawn.  $\times 500$ .

tive effects of the two heat treatments upon the contraction in area, elastic limit and resistance to impact can be clearly seen by comparing the

Steel No.	Heat treatment	Results of the mechanical tests				
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.	Impact resistance (Charpy test) kg.-m. per sq. cm.
1	Annealed.....	92,900	54,500	18.2	48.2	12.3
	Quenched and drawn.....	100,800	85,500	18.6	62.8	22.2
2	Annealed.....	97,400	57,300	20.0	43.2	13.1
	Quenched and drawn.....	115,600	84,500	18.5	65.8	21.8
3	Annealed.....	94,700	56,200	19.1	49.8	9.8
	Quenched and drawn.....	112,800	86,900	17.9	68.7	19.3
4	Annealed.....	94,300	55,000	20.2	41.0	12.1
	Quenched and drawn.....	116,200	88,400	19.1	66.6	20.7

191. Characteristic effects of the two treatments upon the elastic limit are still more evident when the chemical composition is such that the steel may be manipulated so as to obtain nearly constant tensile strength, while maintaining all or at least a good part of the efficiency of the heat treatment.

The following table contains tensile properties of three such steels containing about 0.45 per cent. carbon, 0.50 per cent. chromium, 0.50 per cent. nickel and 0.60 per cent. manganese. Annealings were made between 800°C. and 850°C.; quenchings after an equal heating were in water at about 20°C. Drawing the quenched material was done between 600°C. and 630°C. for a length of time sufficient to bring the tensile strength back to a value near that of the same steel simply annealed.

Usual test-pieces were obtained, cut exclusively in a longitudinal direction from forged bars subjected to the specified heat treatment.

Steel No.	Heat treatment	Results of the mechanical tests		
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.
1	Annealed.....	107,400	59,700	15.5
	Quenched and drawn.....	103,800	78,200	17.0

As can easily be seen, large differences remain in the elastic limit of each steel even when the corresponding differences in the breaking load are very small.

The facts indicated in subdivision (a) of Sec. 189 have such an important practical bearing that I think it well to illustrate them further with another example taken from routine operations.

A hollow forging weighing 16 tons was obtained from an ingot of medium hard 2 per cent. nickel steel weighing about 50 tons, forged upon a mandrel. The walls of the hollow cylinder after rough turning and boring had a thickness of about 170 mm.

Test-pieces for standard mechanical tests (specified in Sec. 151) were taken from about the middle of the wall, and in such a way that their principal axis was parallel to the tangent of the corresponding straight section of the cylinder. Considering the process of forging it is clear that these test-pieces must be considered as *longitudinal* test-pieces, in the sense being parallel to the elongation produced by forging.

Tests made upon the piece after annealing for 6 hours at 850°C. and left to cool slowly in the furnace gave the following results:

Tensile strength.....	87,500 lb. per square inch
Elastic limit.....	44,100 lb. per square inch
Elongation.....	20 per cent.
Reduction of area.....	33 per cent.
Resistance to impact (Charpy test)....	6 kg.-m. per square centimeter

The structure of the metal thus treated, reproduced at an enlargement of 100 dia. in Fig. 188, shows the heavy segregation of ferrite and pearlite, produced during the very slow cooling of the piece.

The same piece was heated again at 800°C. during 2 hours and then chilled in the open air.

Tests made upon the piece thus treated gave the following results:

Tensile strength.....	90,200	lb. per square inch
Elastic limit.....	48,400	lb. per square inch
Elongation.....	23	per cent.
Reduction of area.....	44	per cent.
Resistance to impact (Charpy test)...	7.15 kg.-m.	per square centimeter

Microstructure of the latter at 100 dia. (Fig. 189) is characterized by a great deal smaller dimension of the crystalline elements. In the distribution of the structural elements, however, there remains a trace of the

mixed in about equal parts.



FIG. 188.—Large crystals of ferrite in annealed gun tube.  $\times 100$ .



FIG. 189.—Microstructure of gun tube of Fig. 188 after air quenching.  $\times$

Finally, the same piece was again heated at  $900^{\circ}\text{C}$ . for 4 hours, left to cool slowly in the furnace to  $800^{\circ}\text{C}$ . and then quenched in water. Heat treatment was completed with a drawing at  $640^{\circ}\text{C}$ .

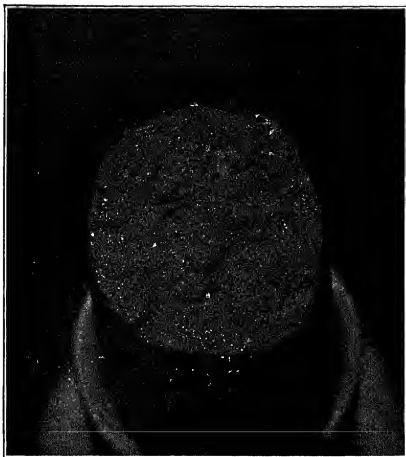
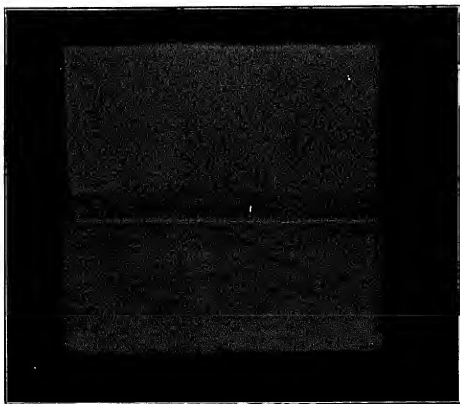


FIG. 190.—Tension fracture, gun tube after air quenching.  $\times 5$ .





Results of comparative mechanical tests made upon the metal thus treated were as follows:

Tensile strength.....	105,700	lb. per square inch
Elastic limit.....	66,800	lb. per square inch
Elongation.....	26	per cent.
Reduction of area.....	57.3	per cent.
Resistance to impact (Charpy test) ..	14.2	kg.-m. per square centimeter



FIG. 192.—Gun tube, water quenched and drawn.  $\times 100$ .

The structure of the quenched metal after being drawn is reproduced at 100 dia. in Fig. 192. It is very evident that the final heating has permitted only a partial segregation of the ferrite; the proportion of free ferrite in fact is a great deal less than after the two previous treatments (compare Fig. 188 and 189). Therefore, we may conclude that the last heating produces results intermediate between those of a drawing (or tempering) and those of an annealing, properly speaking. It is interesting to observe that almost all the ferrite has developed in the lamellar forms which we have demonstrated to be characteristic of separation from quite homogeneous solid solutions directly into the  $\alpha$  state. Therefore, microstructure is also fully in accord with mechanical tests in confirming the efficacy of this particular quenching for homogeneity.

Tension and Charpy impact fractures (reproduced respectively in Fig. 193 at about five times and in Fig. 194 about twice the real size) also fully confirm the same conclusions. In fact both the surfaces have a finely fibrous structure throughout almost uniform in texture.



dimensions in the open air, may give rise to much different effects from those caused by a true<sup>1</sup> annealing.

This is especially evident when it is observed that the effects of the second treatment mentioned in Sec. 192 are similar to, but greater than, those of the first. Therefore, if, on account of the slowness of cooling due to the large dimensions of the piece, the second heat treatment had exhibited the features of a true annealing, it should have caused a *lesser* value for tensile strength and elastic limit; or at least, presenting the hypothesis least favorable to our argument, it should not have increased those values over the results from the first experiment. The fact that none of those two situations occurred, but instead a substantial increase of the tensile strength and of the elastic limit, proves that the observed effects are due to a true quenching process, although extremely mild and *attenuated*. We have seen that this conclusion is confirmed by the microstructure.

For the reasons already explained, air chilling, quenching in molten lead, and other mild quenchings when utilized as *homogeneity* heat treatments, may affect the material far more profoundly when applied to special steels characterized by very low transformation points upon cooling. This is even more true of forged or rolled steels than of special steel castings, and consequently assumes a greater practical importance, a fact which is easily comprehended when one remembers that steels subjected to plastic deformation are a great deal more sensitive to homogeneity heat treatments, for various reasons indicated in Chapter XIX.

The practical importance of these facts consists in this: that the use of soft and medium alloy steels possessing the distinctive properties previously indicated permits one to attain excellent results on intricate forms or large dimensioned pieces which could not be subjected to the shock of energetic quenching without serious trouble due to warpage, checks, cracks and the like. In steels of such analyses, even the mildest quenching produces microstructures and improvement in elastic limit, contraction and impact strength fully as good as the most drastic cooling and drawing would produce in plain carbon steels.

It is seen that we are in command of another method by which it is even possible to obtain results similar to those described in Sec. 188, reached by low-temperature quenching of steels possessing a strong thermal hysteresis.

Besides, the ideas underlying the two methods are clearly similar.

**194.** It is opportune to give at least one example of the phenomena mentioned in Sec. 193 and of the processes derived from their commercial application.

<sup>1</sup> That is to say, a long reheating at a temperature above the transformation range, followed by a very slow cooling.

A steel containing about 0.25 per cent. carbon, 0.80 per cent. chromium and 2.80 per cent. nickel, was made in the acid open-hearth furnace and cast in 550 × 550 mm. square ingots, which were afterwards forged under a press to flat blooms about 200 × 300 mm. in cross section.

The following table contains a description of the various heat treatments to which four of these blooms were subjected and the results of tensile tests made upon the usual longitudinal test-pieces.

It should be remembered that in this case, as well as in the tests mentioned in Sec. 192, the coupons were cut from the central or deepseated portion of each bar; namely, from the region where the cooling velocity was necessarily moderate.

No.	Heat treatment	Results of tension tests			
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.
1	Heated at 850°C. for 6 hours, followed by slow cooling in the furnace.....	94,400	56,900	21	42
2	Heated as above, followed by quenching in oil at 30°C.....	105,500	79,600	22	58
3	Heated as No. 1 followed by slow cooling to 700°C. and then by air chilling	107,400	78,210	21	56
4	Treatment as No. 3 and then drawn at 600°C. for 2 hours.....	108,300	76,800	20	55

Number 3 shows that air chilling this steel gives practically identical results as when quenching in oil (No. 2), while these results differ substantially from those obtained by a true annealing (No. 1) especially as far as elastic limit and reduction of area are concerned.

Data concerning test-piece No. 4 show that drawing this steel at 600°C. does not modify the mechanical properties substantially, because the differences between the values given by No. 4 and No. 3 are within the limits of experimental error. Therefore, we now have one of those cases which I have mentioned in another connection, in which an "attenuated quenching" (in this case in oil or in air) constitutes a complete heat treatment by itself, comprising in one operation the effects of normalizing and of a final heat treatment for quality. It has already been mentioned that in these cases a final drawing may sometimes be advisable (although it does not modify the mechanical properties imparted to the steel by the previous treatment) simply as a means to eliminate any internal stresses produced during the mechanical work or during the heat treatment.

## CHAPTER XXII

### COMPARATIVE EFFECT OF ANNEALING AND QUENCHING UPON THE TRANSVERSE STRENGTH OF HOT WORKED STEEL

195. Let us now see some examples showing the characters of the phenomena mentioned briefly in Sec. 189(b). They comprise the second group of main practical advantages obtainable by homogeneity quenchings, as compared to the results made possible by a simple annealing.

Essentially the advantage in quenching rests in the better suppression of all those characters which together are responsible for the transverse weakness occurring in greater or lesser measure in all forged or rolled steels.

It is necessary to remind the reader of the observation already made in another connection: Generally, transverse and longitudinal properties do not differ greatly in very pure and well deoxidized steels, while serious and dangerous transverse weakness occurs in greater measure with increasing degree of oxidation at the time of solidification. Emulsified sonims are always contained in oxidized steels although their dimensions might be so small as to render them hardly visible under microscopic observation made without special care; yet when they appear in exceptionally great quantities, transverse weakness is sure to be pronounced.

It follows that a practical study of transverse weakness in forged or rolled steels may be conducted at the same time one considers the effects produced upon them by emulsified and more or less oxidized sonims.

In mentioning the experiments by which I propose to sketch the characters of transverse weakness, I shall follow this criterion exactly. Therefore, I shall begin by describing some very pure and well deoxidized steels, nearly isotropic, and shall then pass to highly oxidized steels containing many strongly oxidized emulsified inclusions.

196. First of all it is opportune to show by a concrete example that in some classes of work it is necessary to use the greatest care that transverse weakness does not exceed a certain value in relation to the stresses to be withstood by each individual part of the machine, and to study carefully the various aspects under which the defects may appear. In this way, the practical importance of the forthcoming data will be more evident. It will also be clearer that from this special point of view, the technical problems connected with mechanical working and with homogeneity heat treatment of machine steels must be studied with great care.

Figure 195 reproduces at about two-thirds actual size a sulphur print taken upon the surface obtained by cutting the neck of a crank-shaft for a

six-cylinder aeroplane motor along a plane passing through the axes of the adjacent bearings. Sulphur printing, as is perhaps well known, is done by pressing a sheet of silver bromide photographic paper moistened with dilute sulphuric acid upon the polished surface of the steel for a few seconds; the sulphuric acid, in attacking metal and its inclusions reacts with the formation of hydrogen, mixed with a small proportion of hydrogen sul-

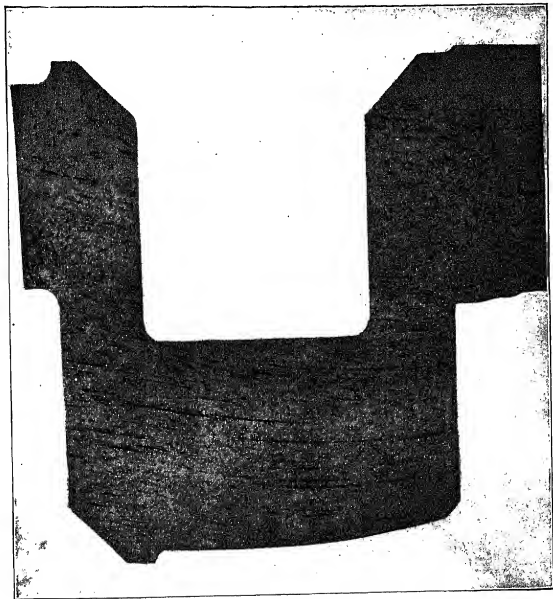


FIG. 195.—Sulphur print of longitudinal section of crank shaft.  $\times \frac{3}{4}$ .

phide, a proportion which is greater the higher the percentage of sulphur in the steel. The  $H_2S$  thus formed, being in close contact with the silver bromide at the points where it is evolved, is immediately fixed in location by reacting to form silver sulphide. The sensitive paper is therefore tinted in its various points by a color which is denser the higher the sulphur concentration in the corresponding part of the metallic surface.

corresponding to the periphery of the primary crystals of steel (as we know from the discussion given in Parts I and II), the distribution of the darker regions of a sulphur print will furnish a picture of the form and distribution of the structural elements derived from the  $\gamma$  mixed crystals.

A roughly parallel alignment of these structural elements appears clearly in the case represented by Fig. 195. This alignment is about the same for the various parts of the piece under examination, which is due to the method of manufacture. A flat bar was first produced under the hammer which was then shaped into three rectangular blocks connected at their centers by a neck of steel later to be turned to fit the intermediate bearings. From these square blocks were then machined the finished crank hangers and the piston pin. Observation of Fig. 195 confirms the conclusion evident from this description that the two throws of the crank shaft will be subjected to bending stresses in service in a direction transverse to the principal direction of forging. That is to say, the shaft will be loaded in a direction in which the effects of transverse weakness are felt in the most dangerous and pronounced way.

It is clear that in cases similar to the one just described it is necessary to reduce this transverse weakness to the extreme minimum by means of special precautions during melting and pouring of the steel, and by means of a rational study and very accurate application of heat treatments for homogeneity.

When the melting and heat treating practice were not so well understood, so that approximate equality in physical properties in all directions could not be attained, especially in the high tensile steels, it was necessary in many cases to modify the forging practice—even by greatly complicating the manufacturing processes—so that the parts of the member which was to withstand the greater stresses would eventually be formed of material which had been forged longitudinally to the direction of said stresses. Thus, in cases similar to the one illustrated in Fig. 195, it was customary to start by forging a straight bar with diameter a little larger than the greatest diagonal of the finished square section of the crank hanger. The bar was then bent, often in expensive dies, into the form of the completed shaft. Again, whenever hollow cylindrical bodies designed to withstand tangential stresses were to be manufactured, as, for instance, rotors for high velocity steam turbines, guns, tubes, jackets, etc., recourse was made to forging upon a mandrel, even for pieces of smaller diameter. At the present time this process, very difficult and ex-

ments, primarily on account of steel well deoxidized when poured and containing a low proportion of emulsified sonims.

The first of these cases is the large hollow cylindrical piece described in Sec. 192 made of 2-per cent. nickel steel and obtained by forging upon a mandrel. Due to the method of working, the radial direction is clearly transverse as regards direction of elongation.

Therefore, I shall indicate the results of the physical tests made upon heat treated test-specimens of the dimensions indicated in Sec. 151 so cut that their axis corresponds to a radius of the cylinder. The three heat treatments are as indicated in Sec. 192.

After annealing at 850°C. for 6 hours the results were the following:

Tensile strength.....	84,800 lb. per square inch
Elastic limit.....	44,100 lb. per square inch
Elongation.....	16 per cent.
Reduction of area.....	18 per cent.
Resistance to impact (Charpy test)....	4 kg.-m. per square centimeter

After the second treatment (air chilling after heating at 800°C. for 2 hours) the results were as follows:

Tensile strength.....	92,200 lb. per square inch
Elastic limit.....	49,800 lb. per square inch
Elongation.....	15 per cent.
Reduction of area.....	16 per cent.
Resistance to impact (Charpy test)....	5 kg.-m. per square centimeter

Finally, after the third treatment (heating at 900°C. for 4 hours followed by the slow cooling to 800°C.; then quenching in water and finally, drawing at 640°C.) tests gave the following results:

Tensile strength.....	104,800 lb. per square inch
Elastic limit.....	65,400 lb. per square inch
Elongation.....	22 per cent.
Reduction of area.....	48.5 per cent.
Resistance to impact (Charpy test)...	12.8 kg.-m. per square centimeter

Comparison of the data in the above three groups between each other and with the corresponding ones reported in Sec. 192 from *longitudinal* test-pieces, suggests the following observations:

This annealed steel is but slightly deficient in transverse properties, proving that the steel used was remarkably pure and well deoxidized. Appreciable deficiencies appear only in the figures for reduction of area and resistance to impact.

In the air-cooled steel (a mild quenching which, due to the dimensions of the piece, is very nearly a true annealing) the elongation and contraction are practically unimproved, while there is a slight gain in the resistance to impact. Altogether it can be observed that the improve-



annealing, is substantially less in the transverse than in the longitudinal tests.

Finally, in the steel heated at 900°C. quenched and drawn, the improvement in the transverse physical properties is a great deal better than in the longitudinal; so much so that one may consider that this heat treatment has entirely eliminated the effects of any pre-existing transverse weakness.

Here therefore is another example showing still more clearly that the efficiency of homogeneity quenchings is much higher than the corresponding annealings.

**198.** As has been already pointed out, and after all is evident from the arguments heretofore developed, the indications and quantitative measures of transverse weakness vary within quite large limits, starting with very pure and perfectly deoxidized steels in which the effects are of small amount and can be eliminated nearly completely by proper heat treatment, and ending with some impure steels, strongly oxidized and strewn with abundant non-metallic inclusions, steels so weak transversely as to render the metal really unfit for the greater part of its intended uses. Such extreme cases are little improved by the most energetic heat treatment.

One extreme of this series has already been illustrated immediately above. Now before describing some other more difficult cases, I think it best to describe a series of simple tension experiments, made upon steels of similar composition and forged under the same conditions, arranged according to the increasing amount of transverse weakness which they exhibit. In this way it will be easier for the reader to form a synthetic idea of the limits within which the phenomenon may vary due simply to variations in purity.

In order to better emphasize the anisotropy of each heat, and in order to confirm the fact that even the worst of them may possess very satisfactory mechanical properties when determined in the longitudinal direction, I shall also give the results of longitudinal tension testpieces.

The steels were made by the acid open-hearth process and cast in octagonal ingots weighing 12 to 15 tons. They were all medium-hard 2-per cent. nickel steels. The ingots were forged under the hydraulic press and reduced to cylindrical bars of about 350 to 400 mm. dia. The bars then were rough turned, bored and subjected to standardized annealings and quenchings for homogeneity, made in accordance with the criteria already mentioned. These preliminary heat treatments were followed by simple drawings, designed to bring the tensile strength and the elastic limit to the desired values.

The transverse and longitudinal tests were made under identical

No.	Transverse tests					Longitudinal tests				In bending
	In tension				In bending	In tension				
	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.	
1	105,300	78,200	18	48	OK	101,000	75,400	19.5	52.0	OK
2	115,200	81,100	19	43	OK	118,100	79,600	20.0	55.0	OK
3	107,400	82,500	20	50	OK	109,500	81,100	22.0	61.0	OK
4	110,400	82,800	14	33	OK	112,200	76,100	23.0	52.0	OK
5	109,400	74,200	13	30	OK	106,700	75,400	21.5	48.0	OK
6	113,200	87,500	12	28	OK	110,900	82,500	22.2	62.0	OK
7	110,000	75,100	9	5	cracked broke at 80° cracked	107,400	91,000	19.8	49.0	OK
8	114,400	84,700	7	11		116,300	80,200	20.2	47.4	OK
9	116,100	85,600	8	14		118,100	94,200	19.6	46.4	slightly cracked
10	114,600	85,300	5	9	broke at 98° broke at 120° broke at 150°	112,200	75,100	20.0	38.9	slightly cracked
11	91,500	74,500	3	0		114,200	89,600	18.0	43.5	OK
12	82,500	40,800	1	0		118,300	83,900	19.0	48.2	slightly cracked

This table requires few comments. The data are gathered in four groups, the first of which includes heat treated steels which are nearly isotropic in physical properties. The succeeding groups contain data relative to steels in which the dangerous transverse weakness appears with an ever increasing intensity in spite of the heat treatment. The three steels of each group are about equally deficient in toughness across the grain.

199. That homogeneity heat treatments eliminate, or at least suppress, transverse weakness is even more evident when using a more energetic heating and cooling program (as we shall see further on by some concrete example) or when a definite treatment is repeated.

In general it is a fact that the efficacy of a preliminary heat treatment when based upon the use of quick coolings is very much greater when more than one quenching is made, than when the heating of the same degree and total duration is made continuously in one phase, and followed by only one quenching.

Steel No.	Heat treatment No.	Description of repetitive heat treatment	In tension				In bending
			Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction, per cent.	
1	I	Preliminary quench—followed by a draw.	106,400	83,600	8	14.3	Broke at 90° Cracked
	II	Two preliminary quenchings (identical to that of No. 1) followed by a drawing.	107,400	74,500	13	24.2	
	III	Three preliminary quenchings (identical to No. 1) followed by a drawing.	105,300	74,700	20	37.5	OK
2	I	Preliminary quench followed by a draw.	121,500	86,800	6	18.0	Broke at 110° Broke at 90°
	II	Two preliminary quenchings (identical to that of No. 1) followed by a drawing.	118,500	83,100	10	10.0	
	III	Three preliminary quenchings (identical to No. 1) followed by a drawing.	115,900	85,600	15	26.0	Cracked
	IV	Four preliminary quenchings (identical to No. 1) followed by a drawing.	112,200	78,900	19	34.0	OK

Instead it is more desirable to give some experimental data showing the enhanced effects obtained by several repetitions of a given heat treatment, in comparison with those that the *same treatment* produces when applied only once, or a smaller number of times.

The figures contained in the above table, refer to steels of the same type as those mentioned in the previous section, manufactured, treated and tested under the same conditions. In this case, I deem it superfluous to add the results of longitudinal tests. Neither do the results require any comment.

200. Finally, it sometimes happens that steel of a certain heat appears "refractory" against normalization, at least as far as concerns an attempt to eliminate or mitigate the effects of transverse weakness.

Two examples of this phenomenon are shown in the table on page 337, containing the results of testing 2-per cent. nickel steel forgings made under the same conditions as those described in the two previous sections.

In the greatest majority of the cases in which the steel is exceptionally weak when tested across the grain, and in which this defect cannot readily be eliminated by treatments ordinarily efficacious, such a condition of affairs exists in all the forged or rolled pieces obtained from ingots of that heat. This is a clear indication that the trouble originates from the nature of the raw materials charged into the furnace, or from the conditions under which the melting, refining and deoxidizing operations have been carried out. Other conditions being equal, experience demonstrates

Steel No.	Heat treatment No.	Description of repetitive heat treatment	Transverse tests					Longitudinal tests				
			In tension				In bending	In tension				In bending
			Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.	
1	I	One homogeneity quenching from 850°C., then drawn at 600°C.	110,900	72,500	9	7	Cracked	106,700	82,500	19	39	OK
	II	Two homogeneity quenchings from 850°C., then drawn at 600°C.	118,100	81,100	8	6	Cracked	110,900	79,600	18	45	OK
	III	Four homogeneity quenchings from 850°C., then drawn at 600°C.	113,800	71,100	9	7	Cracked	110,900	82,500	21	41	OK
2	I	One homogeneity quenching from 900°C., then drawn at 600°C.	118,100	86,800	7	7	Broke at 120°	119,500	83,900	15	35	OK
	II	Two homogeneity quenchings from 900°C., then drawn at 600°C.	120,900	93,900	6	12	Broke at 120°	118,100	83,900	14	26	OK
	III	Four homogeneity quenchings from 900°C., then drawn at 600°C.	120,900	91,000	6	9	Broke at 120°	120,900	83,900	15	31	OK

that the deoxidizing operations are those which if unskillfully conducted will leave the most serious after effects.

However, there are cases in which the phenomenon of transverse weakness appears in quite different degree and persistency in pieces obtained from the different ingots of the same heat, and sometimes even in different parts of the same ingot, or in a single piece obtained from one portion of an ingot.

It is clear that, in these last cases, the defect must originate during teeming or during the metal's solidification.

This hypothesis is very frequently confirmed by microscopical analysis, revealing the presence of emulsified sonims. From their position in relation to the crystalline ferrite masses, these inclusions appear to be strongly oxidizing in their influence, in accordance with what was said in Sec. 97 and 98. Now a great many observations demonstrate that emulsified sonims of certain types may mix with the steel in very different degree during different phases of the teeming, and thus remain suspended in the metal in different proportion in various parts of the same ingot or of the same casting.<sup>1</sup> I cannot dwell upon the reasons for this segregation, but inasmuch as transverse weakness may be strongly accentuated by the presence of emulsified oxidized inclusions, the same reasons which explain the irregular distribution of those inclusions into different ingots of the same heat or in the different parts of the same ingot or casting will also explain the irregular results of physical tests.

As an example, a steel of the following composition:

Carbon.....	0.30 per cent
Manganese.....	0.62 per cent.
Silicon.....	0.26 per cent.
Sulphur.....	0.02 per cent.
Phosphorus.....	0.06 per cent.

made in a 35-ton acid open-hearth furnace, was cast in octagonal ingots weighing about 7 tons each. The ingots were forged by a press into round bars about 300 mm. in diameter. After rough turning, the bars were subjected to a heat treatment for homogeneity. Those obtained from the first two ingots cast showed only a slight transverse weakness, easily eliminated by normalizing. On the other hand, the bars obtained from the last three ingots were very brittle transversely, and they were improved but little and with difficulty by the heat treatments.

tained from the first ingot cast, after forging and annealing for 6 hours at 850°C. followed by slow cooling. The steel contains few non-metallic inclusions, and these do not appear to have a strongly oxidizing influence. In fact, the various non-metallic particles to be seen in the figure are not surrounded by particularly well developed ferrite. On the other hand, some, as for instance the inclusion at the lower left side of the figure, and the long one appearing at the right, are actually in contact with pearlite.

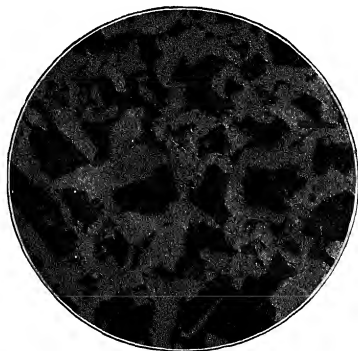


FIG. 196.—Microstructure of 0.30-per cent. carbon steel forging containing harmless inclusions.  $\times 100$ .

A transverse tensile test upon the steel thus treated gave the following results:

Tensile strength.....	81,500 lb. per square inch
Elongation.....	22 per cent.
Reduction of area.....	26 per cent.

The same bar, annealed again for 6 hours at 780°C. and left to cool in the air, showed the structure reproduced in Fig. 197 (enlarged 100 dia.) and gave the following results when testing transversely:

Tensile strength.....	87,500 lb. per square inch
Elongation.....	21 per cent.
Reduction of area.....	38 per cent.

Aside from the reasons for the small dimensions of the crystals of ferrite, which we know, we could repeat for Fig. 197 the same observations made about the structure of Fig. 196. The mechanical properties in the two cases also prove that the transverse properties are nearly equal to the longitudinal, especially after the mild quenching.

The steel in the fifth ingot of the same heat was quite different.

Figure 198 reproduces at an enlargement of 100 dia. the structure of the bar obtained by forging in the same manner as described for the



FIG. 197.—Same steel as Fig. 196 after air quenching.  $\times 100$ .



FIG. 198.—Banded ferrite around oxidized inclusions.  $\times 100$ .

ingot, followed by an anneal at  $850^{\circ}\text{C}$ . for 6 hours. It was cooled slowly in the furnace.

The inclusions are more abundant in this bar than in the first

masses of ferrite which have gathered around them, forming lamellar masses appearing in section as bands. They are the principal cause of transverse weakness.

Tension tests made upon this bar confirm the expectations, giving results as follows:

Tensile strength.....	82,500 lb. per square inch
Elongation.....	13 per cent.
Reduction of area.....	6 per cent.

Transverse tests made upon the same bar after having been subjected to a second heat treatment (annealing at 780°C. for 2 hours followed by cooling in the air) gave the following results:

Tensile strength.....	88,200 lb. per square inch
Elongation.....	12 per cent.
Reduction of area.....	11 per cent.



FIG. 199.—Same steel as Fig. 198 after air quenching.  $\times 100$ .

As seen, the physical properties remain relatively unaltered even after the second heat treatment, which also leaves almost entirely untouched the masses of ferrite gathered around the inclusions. This fact is shown in Fig. 199, ( $\times 100$ ) representing the microstructure of the bar taken from the fifth ingot, after having been subjected to the second treatment.]

201. As has been pointed out, it sometimes happens that different parts of the same piece, often very near each other, give quite differ-



In some cases microscopic examination of the metal immediately establishes the fact that these differences are due to a different amount of solid non-metallic inclusions in the various parts of the piece; as an instance Sec. 106 may again be consulted.

But it may also happen (as mentioned in Sec. 173) that very great differences in transverse properties may occur in the various parts of a forged or rolled steel piece even without being possible to detect noteworthy quantities of sonims with the ordinary means used in microscopy.

A typical example of this follows. Part of an octagonal ingot one meter across flats was forged under the press into the form shown in Fig. 200 and then rough turned. The steel used was a soft steel containing 4 per cent. nickel and 0.16 per cent. carbon.

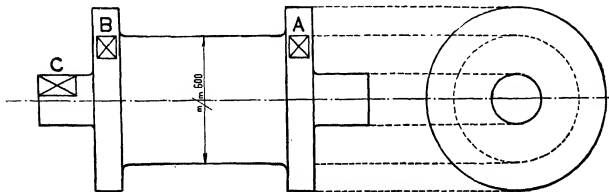


FIG. 200.—Sketch of forging, showing location of test-pieces.

After annealing at 800°C. for 6 hours followed by slow cooling, three longitudinal and three transverse test-pieces were cut at the points indicated in the figure by letters A, B, and C. Tension tests made under the usual conditions gave the results collected in the following table.

Location (Fig. 200)	Longitudinal test-piece				Transverse test-piece			
	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elonga- tion, per cent.	Reduction of area, per cent.	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elonga- tion, per cent.	Reduction of area, per cent.
A	83,600	54,000	18	40	82,800	51,200	4	0
B	80,100	46,900	22	52	77,900	45,500	13	28
C	81,100	46,900	24	63	79,800	48,400	17	25

Flange A evidently shows the phenomenon of transverse weakness very markedly, while it is much less dangerous for test-pieces B and C.

The piece was then heated for 8 hours at 850°C., quenched from 800°C. in water and drawn at 620°C. Transverse test-pieces were again cut from regions A, B and C in a position contiguous to the first ones. Re-

Location (Fig. 200)	Transverse test-piece			
	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.
A	82,500	66,800	3.0	0
B	92,400	69,700	22.0	46
C	91,000	66,800	21.5	38

It appears that quenching has actually decreased the ductility in region *A*, while it has entirely obliterated transverse weakness in region *B*.

Microscopic examination of the metal after this last heat treatment reveals remarkable heterogeneities at region *A* but does not show exceptional quantities of sonims (sec Fig. 201 at 100 dia).

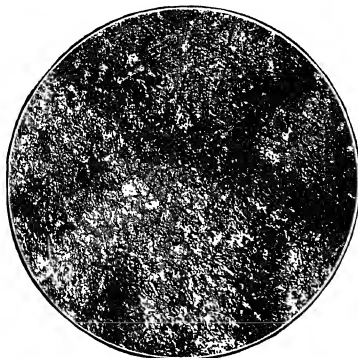


FIG. 201.—Heterogeneity associated with transverse weakness in forging of Fig. 200.  
× 100.

For reasons which we cannot analyze here it appears probable, even when microscopic examination of a polished surface etched with the usual reagents does not reveal the presence of the usual emulsified sonims in abnormal quantities, that the phenomenon of transverse weakness strongly localized in certain parts of a forging is due to an oxidized condition of the metal, associated with the presence of numerous very small non-metallic inclusions in a relation analogous to that observed between the commonly observed emulsified inclusions and the state of oxidation which causes transverse weakness. Such very small particles are observed

only under very high magnifications when examining metallic surfaces carefully polished and not etched.

202. It has already been said at the beginning of Sec. 199 that the efficacy of homogeneity heat treatments may be increased not only by repetition (as has been seen in the last sections) but also by utilization of more drastic or energetic programmes, especially when it is designed to eliminate transverse weakness.

This greater "energy" of normalizing practice must evidently consist in increasing the temperatures to which the various heatings comprising the program are maintained, in prolonging the time at such temperatures, and in increasing the velocity of subsequent cooling. Extremes must of course be avoided, such as would cause various perturbations in the course of processes naturally accompanying normalizing, which deviations have also been noted in the preceding pages.

The following examples are illustrative:

A steel of composition

Carbon.....	0.33 per cent
Manganese.....	0.58 per cent.
Silicon.....	0.33 per cent.
Sulphur.....	0.01 per cent.
Phosphorus.....	0.04 per cent.

was cast in square ingots 550 by 550 mm., and forged under the press into round bars 250 mm. dia.

One piece, after machining, was cut into two equal cylinders which were treated and tested separately. Transverse test-pieces were obtained from the ends of the two cylinders which originally were contiguous in the forging, in order to decrease the probabilities of accidental differences in quality of the metal.

One half-bar was maintained at 820°C. for 2 hours, then quenched immediately in cold water and finally reheated at 700°C. The results of the transverse tension test upon the steel thus treated follow:

Tensile strength.....	81,400 lb. per square inch
Elongation.....	14 per cent.

Figure 202 reproduces at 100 dia. the microstructure of the steel, while Fig. 203 shows the appearance of the fracture. As shown, the break is partly crystalline and partly woody in texture.

The other end of the bar was heated for 6 hours at 900°C., then left to cool slowly in the furnace until it reached 800°C. and quenched in cold water. Finally it was reheated at 700°C.



2.—Microstructure of 0.33-per cent. carbon forging after quenching and reheating.  
X 100.

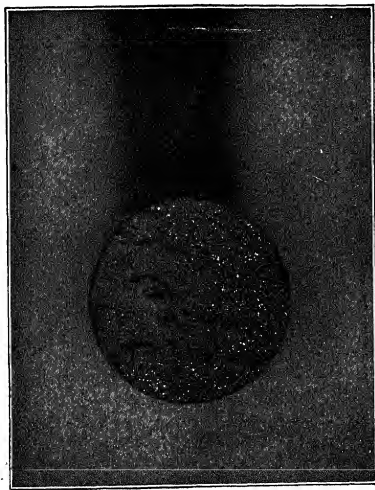




FIG. 204.—Microstructure of steel of Fig. 202 after normalizing, quenching and reheating  $\times 100$ .

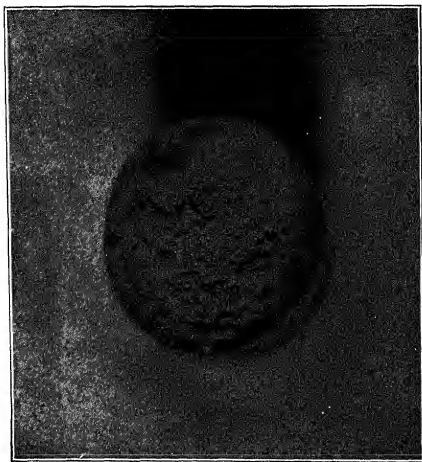


FIG. 205.—Tension fracture of steel of Fig. 204.

Figures 204 ( $\times 100$ ) and 205 reproduce respectively the microstructure and the fracture (now entirely fibrous).

The greater efficacy of the second treatment for eliminating transverse weakness is evident by comparing this data.

**203.** An even more clear and complete example of the fact stated at the beginning of the previous section, is the following:

A steel containing about 0.50 per cent. carbon and 0.50 per cent. chromium was cast and forged in the same way as the one just described.

One of the bars was cut in four half-cylinders by dividing it along two planes, one passing through the axis of the bar and the other normal to the first one. The four parts were subjected to the heat treatments indicated in the second column of the following table. From each one of these, transverse test-pieces were obtained from positions which were contiguous in the whole bar, and when tested under the usual conditions specified in Sec. 151 gave the results tabulated below:

Test-piece No.	Heat treatment	Tensile strength, lb. per sq. in.	Elongation, per cent.	Reduction of area, per cent.
1	Heated for 10 minutes at 800°C.; quenched in water; drawn at 640°C.	99,800	9	4
2	Heated for 30 minutes at 850°C.; quenched in water; drawn at 670°C.	96,700	12	2
3	Heated at 880°C. for 1 hour; quenched in water; drawn at 670°C.	92,400	16	18
4	Heated at 950°C. for 3 hours; cooled slowly to 850°C.; quenched in water; drawn at 670°C.....	101,400	22	47

The greater efficiency appears evident, step by step as the treatment becomes more energetic. But this is still clearer, especially as concerns the most complete elimination of transverse weakness as an effect of the more drastic treatment, when one examines the fracture of the test-pieces.

Thus, Fig. 206 reproduces the break from test-piece No. 2; woody structure is highly developed. We have seen this defect to be characteristic of steels having poor properties when tested at right angles to the

The microstructure of the same test-piece is reproduced at 100 dia. in Fig. 208, showing that we now have a case analogous to that mentioned in Sec. 201; namely, where no inclusions apparently associated with the persistent grouping of ferrite elements into banded masses can be seen upon properly polished and etched sections when examined under moderate magnification, even though the ferrite segregation can be clearly seen.

Finally, Fig. 209 shows the fracture of test-pieces No. 4. The woody structure has now disappeared almost entirely, and is replaced by a fibrous texture. Stratified accumulations of ferrite have also been eliminated

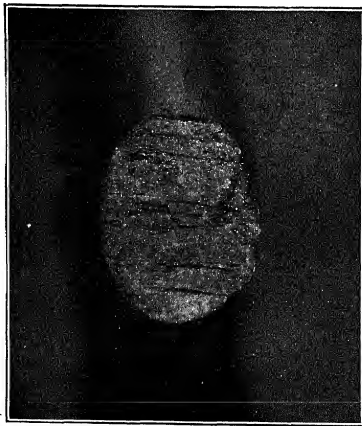


FIG. 206.—Woody structure in test-piece No. 2.

by the energetic heat treatment, as is shown by Fig. 210, which reproduces the microstructure of the same test-piece at an enlargement of 100 dia.

**204.** It sometimes happens that that summation of effects which we have grouped together in the term "transverse weakness" is felt in a very different measure upon the mechanical properties of steel, according to the special property considered. This may happen both for tests made upon longitudinal as well as transverse specimens, the latter, however, being more frequent.

We shall start by considering an example of the first case. In Sec. 160 we have studied the properties of cast metal after various heat

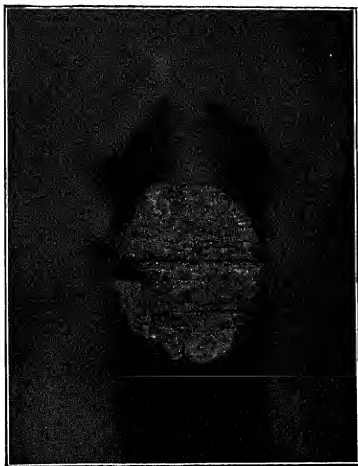


FIG. 207.—Woody structure of test-piece No. 3.

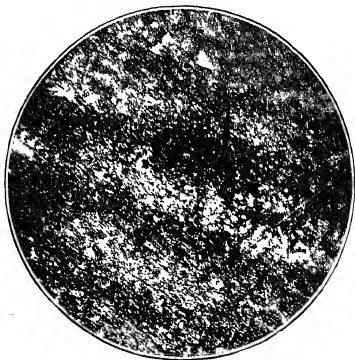


FIG. 208.—Ferrite banding associated with woody fracture of Fig. 207.  $\times 100$ .





FIG. 209.—Fibrous fracture of test-piece No. 4.



FIG. 210.—Microstructure corresponding to fibrous fracture of Fig. 209.

forging and various heat treatments. Inasmuch as we saw in the first place that this steel exhibited ingotism strongly and persistently due to the presence of emulsified oxidized inclusions, we must expect from reasoning advanced at the beginning of this chapter that, after forging, this metal would likewise present marked and persistent transverse weakness.

Experience confirms fully such expectations.

A 220 by 220-mm. square ingot of steel of the same composition as the one indicated in Sec. 160, was pressed into a 170 by 170-mm. square bloom. The cold bar was cut into four parts by sawing it along two planes passing through its axis and normal to the faces. Longitudinal test-pieces were then obtained in such a manner as to avoid the axial region of the ingot, usually the locus of segregated impurities. The successive heat treatments, however, were made upon the entire bar before the corresponding test-pieces were removed.

The following table contains the results of a series of the usual physical tests upon longitudinal test-pieces, obtained after the heat treatments indicated in the second column of the table.

No.	Heat treatment	Tension test			
		Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.	Resistance to impact ( $\rho$ ) (Charpy test), kg.-m. per sq. cm.
1	None; steel as forged.....	105,500	57,600	20	1.7
2	Heated for 4 hours at 800°C. followed by slow cooling.....	103,300	59,000	17	2.5
3	Heated for 4 hours at 900°C. followed by slow cooling.....	101,000	61,900	20	4.1
4	Quench in water from 900°C.; annealed at 600°C. for 4 hr. ....	118,200	74,800	20	5.7
5	Heated at 1100°C. for 4 hours; slowly cooled to 900°C.; quenched in water; annealed at 600°C. for 4 hours.....	123,700	84,300	18	4.9
6	Heated and quenched as No. 5; annealed at 700°C. for 4 hours.....	102,700	64,700	26	6.4

Two transverse bars, one for tension and the other for impact were taken from the original forged bar, treated as indicated in line 4, and

Elongation.....	5 per cent.
Resistance to impact (Charpy test)...	0.5 kg.-m. per square centimeter

These results clearly show the existence of high transverse weakness in accordance with expectations founded upon the behavior of the heat-treated ingot metal.

If now the data obtained in the last table are compared among themselves and with those given in Sec. 160 for the same steel in the unforged condition, and also compared to the elongation and resistance to impact (both in a longitudinal direction) of forged well deoxidized normal steels not infected by overly large quantities of non-metallic inclusions,<sup>1</sup> we see immediately that the improvement in impact produced by forging the impure steel is a great deal less than that caused by a like amount of work upon those other steels, especially when considering the splendid longitudinal tensile strength and elongation acquired by the steel in question by reason of the forging and heat treatment (particularly No. 6 above).

**205.** As I have already pointed out at the beginning of the previous section, the small improvement in one of the physical properties of a transversely weak forged steel, due to a heat treatment which simultaneously produces marked improvement in the other properties of the same steel, may be exhibited in a far larger measure when the experimentation is done with transverse test-pieces.

It is also opportune to indicate an example of this case.

A medium hard steel, contained 0.40 per cent. carbon and 2 per cent. nickel, manufactured in the acid open-hearth furnace, was cast in octagonal ingots weighing about 10 tons each. The ingots were forged down to round bars about 300 mm. in diameter, which were afterwards bored and rough turned, and then subjected to a series of homogeneity quenchings, followed by annealings.

Inasmuch as the tests made upon the first pieces treated revealed high transverse weakness—due in this case to the presence of a great many very small highly oxidized sonims—it was necessary to apply repeated and very drastic homogeneity heat treatments to them before obtaining satisfactory results from static tension tests upon *transverse* pieces. The results of this very numerous series, made under the conditions indicated in Sec. 151, varied between the following limits:

Tensile strength.....	93,900 to 112,400 lb. per square inch.
Elastic limit.....	62,600 to 81,100 lb. per square inch.
Elongation.....	18 to 23 per cent.

Notwithstanding these results, which may be considered as entirely satisfactory, all bending tests were very poor. These were made under the conditions specified in Sec. 151 upon transverse coupons taken im-

<sup>1</sup> See for instance Sec. 179 to 183.

mediately alongside those used for tension. Nor was it possible to obtain an appreciable improvement of these results, even by subjecting the steel to further homogeneity heat treatments. In spite of all efforts, all the test-pieces subjected to bending broke with a strongly woody fracture at low deflections (included angles from  $100^{\circ}$  to  $110^{\circ}$ ).

The contrast between the satisfactory appearance, fibrous fracture, and good necking of test-pieces broken under tension and the woody fracture under bending is well shown in Fig. 211.

**206.** I have already pointed out the most important and probable causes of the fact that the reheatings contained in a more or less complex



FIG. 211.—Bending and tension tests on medium hard nickel steel.

homogeneity heat treatment cannot be prolonged beyond certain limits of time nor pushed beyond a certain limit of temperature without causing overheating. As then noted, the useful results of a slight overheating consist in the more perfect diffusion of the various chemical constituents into the  $\gamma$  solid solution; as the treatment becomes more intense and prolonged the final effect becomes progressively less and less advantageous and afterwards actually dangerous, even ending in a true burning (see Sec. 25, 26 and 83).

It is not necessary to repeat here nor to dwell further upon the origin of phenomena of overheating. It forms an important restriction in all those cases where the most efficient possible homogeneity heat treatments must be applied. On the other hand, some experimental data may be presented which are apt to furnish a precise idea of the effects

overheating might have taken place, I have selected a heating process in which the temperature was always maintained far below the minimum limit necessary to produce a true burning in a steel of that composition. In this case, therefore, the evil effects of over heating ensues only by reason of the *length* of heating.

A low carbon nickel-steel of the following composition was forged in 200 mm. round bars:

Carbon.....	0.20 per cent.
Manganese.....	0.60 per cent.
Silicon.....	0.03 per cent.
Sulphur.....	0.01 per cent.
Phosphorus.....	0.06 per cent.
Nickel.....	2.98 per cent.

It was subjected to the heat treatments noted in the second column of the table below. Inasmuch as these treatments do not include any heatings to a temperature above 1050°C. and inasmuch as a steel of the composition indicated does not pass the liquidus (suffer a true burning) except at temperatures above 1250°C., especially after forging, no doubt may arise but that any dangerous effect produced by thermal manipulation is simply due to long stay at temperatures well above the transformation.

Tests were made upon longitudinal test-pieces.

The effects of overheating are shown by the decrease in elongation and still more by the lessened resistance to impact.

Heat treatment No.	Heat treatment	Tension tests		Resistance to impact ( $\rho$ ) (Charpy test), kg.-m. per sq. cm.
		Tensile strength, lb. per sq. in.	Elongation, per cent.	
1	Heated at 800°C. for 2 hours; quenched in water; drawn at 500°C.....	84,500	30	22.1
2	Heated at 1050°C. for 200 hours; followed by slow cooling in the furnace to 800°C.; quenched in water; reheated at 800°C. for 1 hour, quenched in water; drawn at 500°C.....	86,500	18	11.5
3	Heated at 1050°C. for 400 hours;			

207. Before ending this series of examples, I deem it opportune to present some observations concerning at least one instance not rare in practice, which is difficult to explain by means of the general notions which we possess at the present time and which I have tried to summarize in the previous pages. If nothing else, these observations will suffice to confirm further the fact that our knowledge about the phenomena upon which normalizing processes are based is still very incomplete and imperfect.

In the case which I shall mention, the physical properties of the pieces examined are analogous to those of Sec. 201. In both instances the piece exhibits high transverse weakness in one part while this defect is practically missing in the remaining parts. The steel in question contained 0.44 per cent. carbon, 0.57 per cent. manganese, and 0.50 per cent. chromium. It was cast in 550 by 550 mm. ingots. From the lower half of each ingot, round bars 220 mm. dia. and from five to six meters long were obtained by forging under the press. A piece, 1 to  $1\frac{1}{2}$  m. long, was then cut from each end of the forged bars.

One of such bars, cut, turned, quenched in water after heating for 10 hours at  $850^{\circ}\text{C}$ . and drawn at  $650^{\circ}\text{C}$ . was subjected to transverse tension tests taken from each of its ends, respectively indicated with the letters A and B. The results were as follows:

	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.
End A.....	102,800	74,000	21
End B.....	80,900	46,900	2

Having repeated the same heat treatment and new transverse test-pieces obtained from metal contiguous to that furnishing the preceding ones, tension tests gave the following results:

	Tensile strength, lb. per sq. in.	Elastic limit, lb. per sq. in.	Elongation, per cent.
End A.....	95,700	69,700	19
End B.....	66,700	45,500	2

The most drastic heat treatments, repeated upon the same bar, did not substantially change the mechanical properties of the metal at its two extremities, nor did accurate chemical analysis reveal appreciable differences in composition.

Figure 212 reproduces at an enlargement of 150 dia., the structure at end A after the second heat treatment, while Fig. 213 at the same magnification is the structure of the metal at end B, also after the second heat treatment. These structures were not modified very much by all the subsequent heat treatments.

It was demonstrated in this case that the permanence of high transverse weakness at end B is accompanied by a characteristic structure.

structure which we have examined many times and which usually accompanies to a certain extent the phenomenon of transverse weakness. Crystalline elements delimited by rectilinear boundaries appear even as



FIG. 212.—Microstructure of sound end of forged bar.  $\times 150$ .



FIG. 213.—Characteristic and unexplained microstructure associated with transverse weakness, one meter from Fig. 212.  $\times 150$ .

much more abnormal structures, since they have been observed only in extremely rare cases, and always accompanied by defects in physical properties analogous to those now described. This remark is made as a

result of experience with an enormous number of pieces manufactured in practically the same way as that described for the bar to which the data above mentioned refer, and of steel of practically the same composition. An examination of the few cases analogous to the one indicated which have been discovered appears to justify the opinion that the laws governing the formation and persistence of such crystalline aggregations as shown in Fig. 213 are quite different from those which are summarized in the ordinary equilibrium diagram, or in the general discussion developed in the previous chapters.

As I have already pointed out, this fact constitutes a further proof of the actual imperfection of our knowledge in regard to the phenomena we are now discussing.

**208.** An accurate comparison of all experimental data presented in this last part of our study with those given in Part IV, a comparison which should be conducted under the guidance of the general considerations developed in the first three Parts, would allow one to get even a better idea of the significance and the technical value of such experiments. But inasmuch as this detailed comparison would mean a lengthy exposition, I must permit the reader to work it out for each one of his problems.

If the general considerations developed in this book are well understood, one who attempts such analytical work to solve those of his every day problems as have clearly established fundamental data, will recognize that such work does not present great difficulties. It will then be easy to verify its utility on the basis of the actual results obtained.

As I have already said, the data presented in this study are sufficient so that the great majority of the industrial problems may be analyzed and found to bear upon phenomena none of which are entirely new and such as to confuse the searcher with unforeseen facts. However, I deem it opportune again to call the attention of the reader to two characteristic sides of some observations, doubtless inherent to facts and reasoning already given, yet aspects which must be taken into especial account in the most general application of the ideas which we have developed and illustrated only by particular examples. Inasmuch as I wish to recall the attention to particular groups of ideas already stated, I shall endeavor to be very concise.

**209.** An accurate comparison of the variations caused by certain heat treatments in the physical properties of the types of steels under examination, with the corresponding variations which the same treatments cause in the fracture and the microstructure, establishes the fact that in the greater number of cases the appearance of the surface of fracture cannot be taken as an indication of the state of homogeneity and of



possible under the present state of our knowledge to establish a precise and sure relation between the quality of a steel and its fracture. Nor at any rate could such relation be at all general since we have already seen examples of the fact that the appearance of the fractured surface varies a great deal for the same steel when broken in different ways.

For some particular cases—especially soft and medium cast steels subjected to homogeneity heat treatments<sup>1</sup>—I have had occasion to show how granular crystalline fracture exhibits itself as a consequence of the presence and the development of determined systems of metallographically well defined structural elements, such as a large ferrite network. But the generality of such a co-relation is immediately excluded in the most absolute way by the frequent appearance of coarsely crystalline fractures in steels in which no trace of those microscopic systems exists. In particular, these observations demonstrate that the remarks in Sec. 127 to 129 have a value only for the type of steel there examined, and cannot be applied for instance to steels having very low transformation points, nor to many forged or rolled steels; in fact we know that in these steels the large network of ferrite does not exist, nevertheless they very frequently give coarsely crystalline surfaces of fracture.

Notwithstanding the lack of generality furnished by fracture examinations in the study of actual structure, such inspection may lead to useful practical conclusions whenever it is applied to numerous samples of one particular type of steel, constantly subjected to analogous treatments, especially if this examination is to be utilized essentially as a means of comparison between the results obtained in the various cases after a change in the history of the specimen. A trained eye may then rapidly gather very useful indications even by such an empirical method of observation.

Beside the many examples which involve various appearances of fractured surfaces mentioned in the last two parts of our study, I shall also note here the extended use that it is commonly made of such a method of investigation in the control of armor plate manufacture. In this special branch of metallurgy the maker distinguishes a great number of *types* of fracture, different and well defined. As an example of the subtle distinctions, Fig. 214 shows at about  $\frac{1}{4}$  of its natural size the mediocre structure termed "*fibrous with granular traces*," of an armor plate after it has been forged, cemented, and subjected to the preliminary heat treatment only.

**210.** In Sec. 208 I have observed that the individual reader may apply the principles developed in this book to solve his individual problems. For instance, certain criteria must be borne in mind whenever one must decide whether a simple cast steel (properly heat treated) can be used for

for the heat treatment. This may be done by a comparison of the data given in Parts IV and V.

Now, in the application of such criteria, it is of prime necessity to take careful account of the special effects caused by transverse weakness. In fact we have seen that, if it is true that in many cases normal and constant properties of high tensile strength together with high toughness may be obtained with greater ease in forged or rolled steels by means of homogeneity heat treatments, it is also true that this greater facility is re-

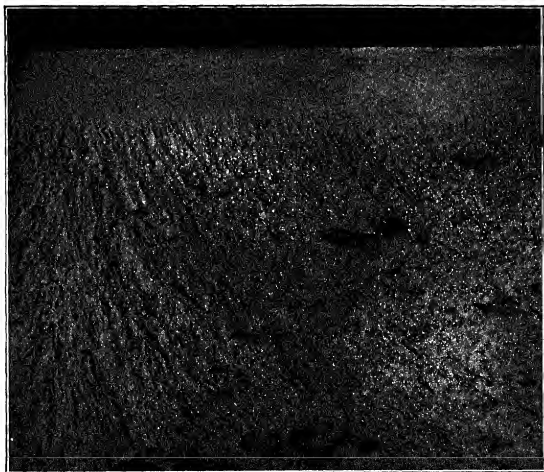


FIG. 214.—Fibrous fracture, with granular traces, of armor plate.  $\times \frac{1}{4}$ .

stricted to the cases in which the mechanical properties are determined by longitudinal tests. On the other hand, when desiring good mechanical properties when tested "across the grain," the data given in the last chapters when compared with those mentioned in Part IV demonstrate that the advantage remains all in favor of cast steels after proper thermal manipulation.

Differences in this regard become still greater when we take account of the probabilities that dangerous accidental irregularities in physical properties may come to light, as described for instance in Sec. 165, due principally to impurities contained in the steel. In fact, while such

irregularities may generally be efficaciously eliminated in forged or rolled steels when tested in a longitudinal direction, they ordinarily appear with high seriousness and persistency, as compared to steel simply cast and heat treated. Naturally this is true if other conditions are equal (especially the purity of the steel) and provided the steels selected are in both cases of suitable composition for the manufacturing processes and treatments to which they must be subjected.

We know that this is mainly due in many cases to the fact that the non-metallic inclusions in the cast steel have a spherical form, while in forged or rolled steel they assume the form of threads or lamellæ acting as *sharp internal notches* and therefore constitute dangerous "primers" for ultimate fracture.

It follows that in many cases, and above all when it is impossible to forge the piece in such a way that the working stresses fall in a longitudinal sense, the rational employment of heat treated steel castings may present relevant advantages in comparison with steel forgings, not only from the viewpoint of rapidity of construction, but also of resistance to stresses and its immunity from accidental fragility.

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